AUGUST 1959

JOURNAL OF THE IRON AND STEEL INSTITUTE

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VOL. 192 PART 4

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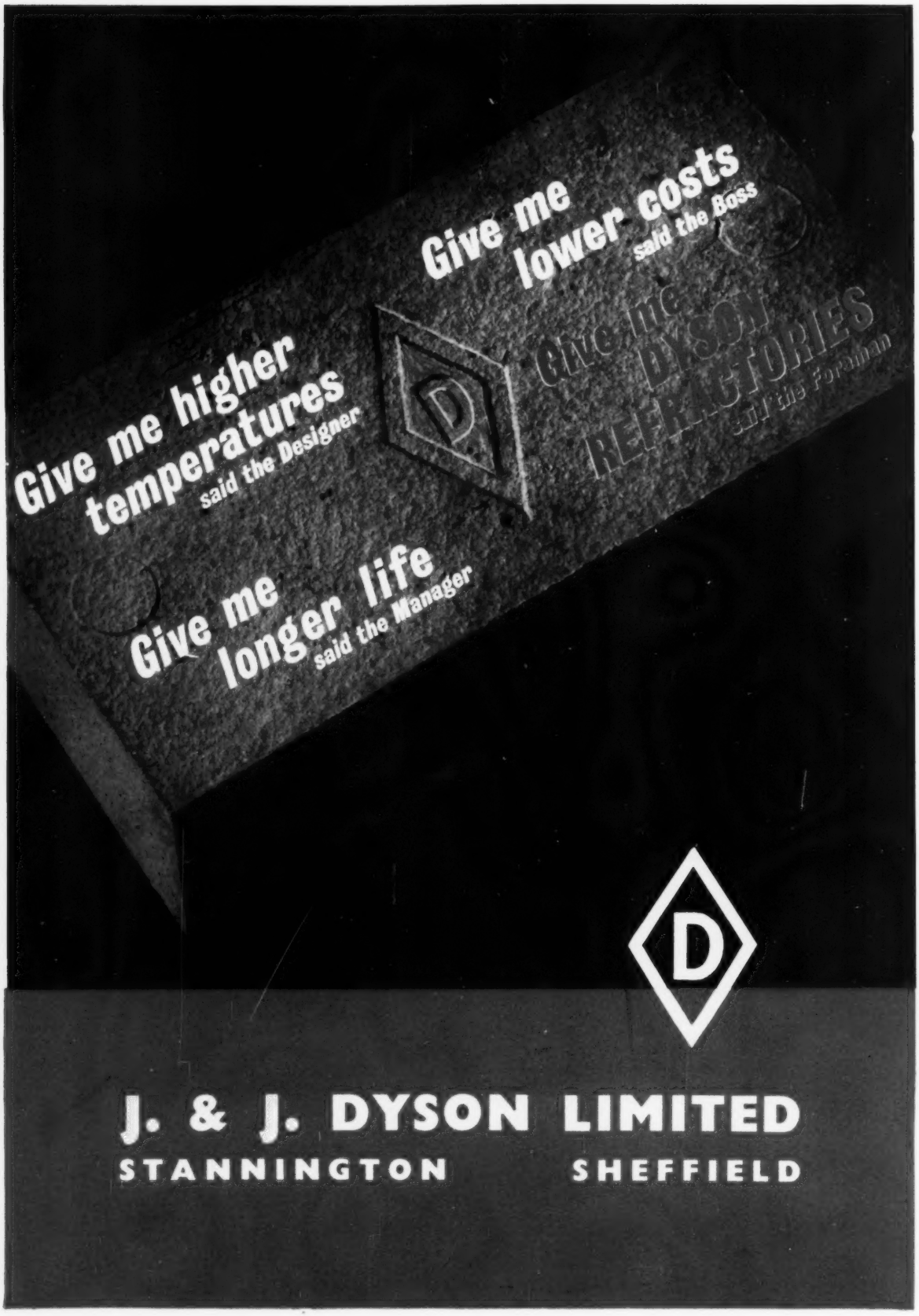
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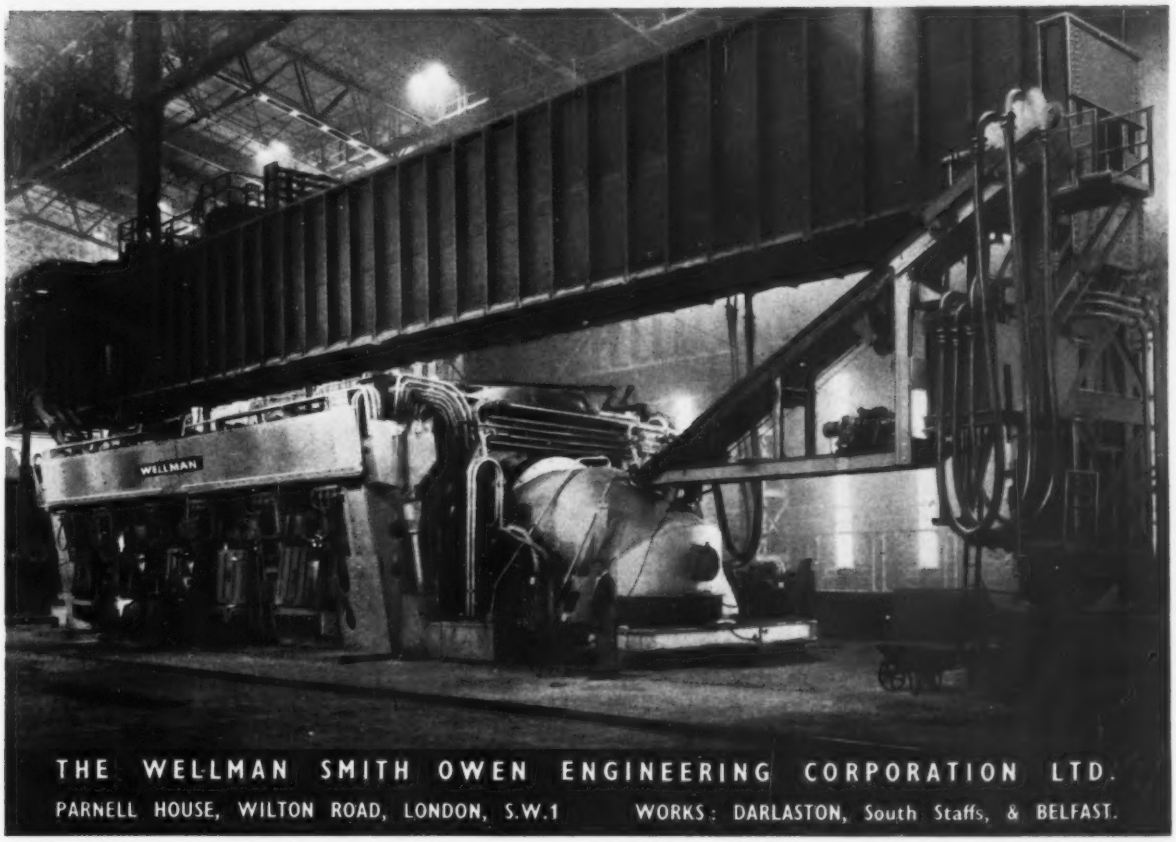

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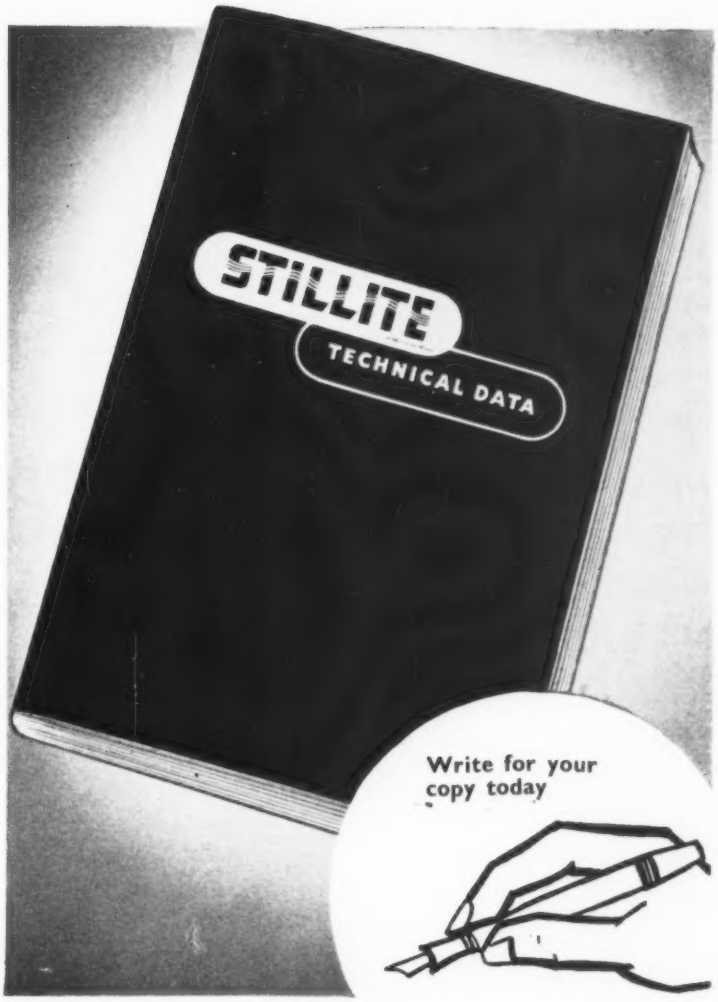

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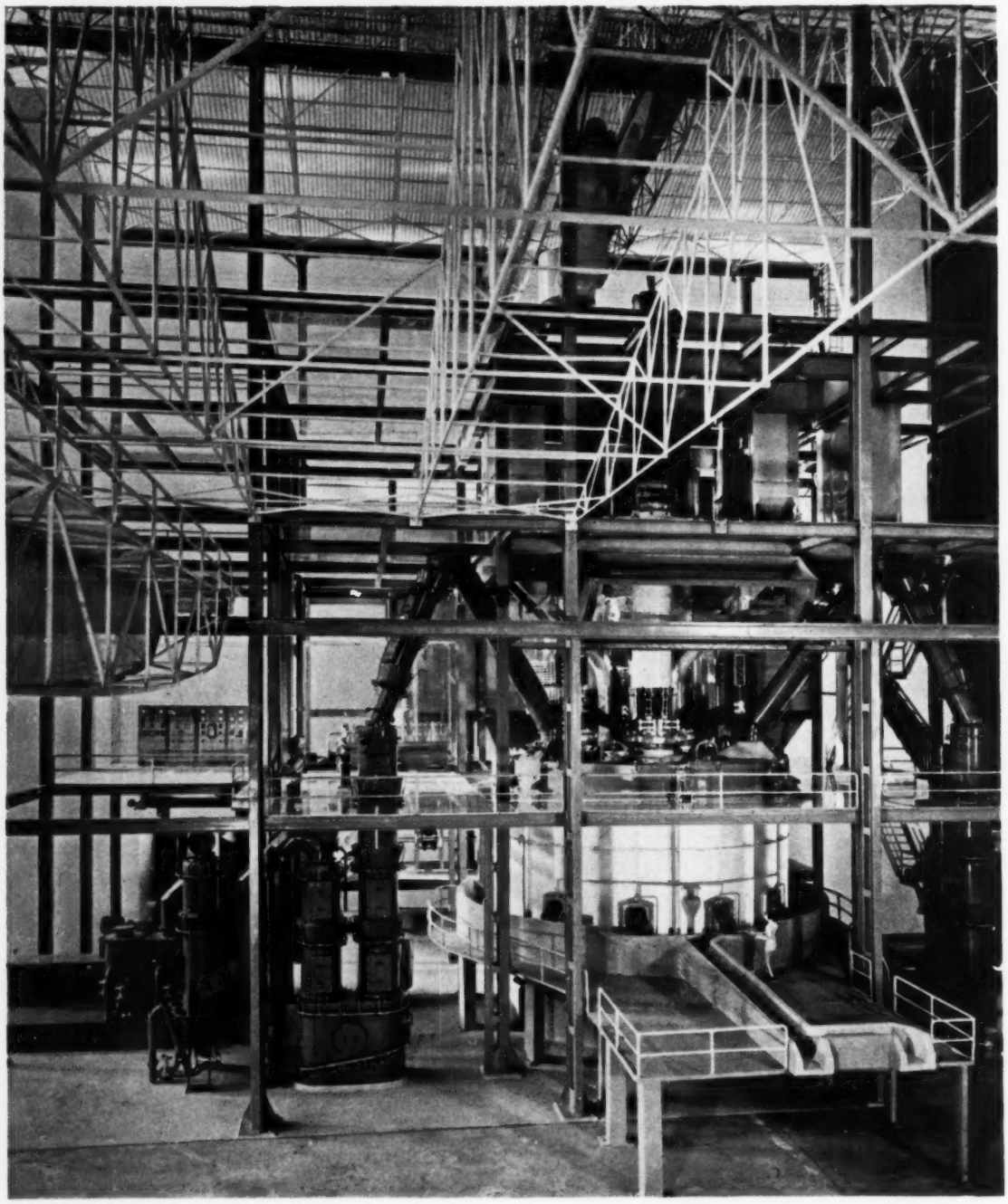

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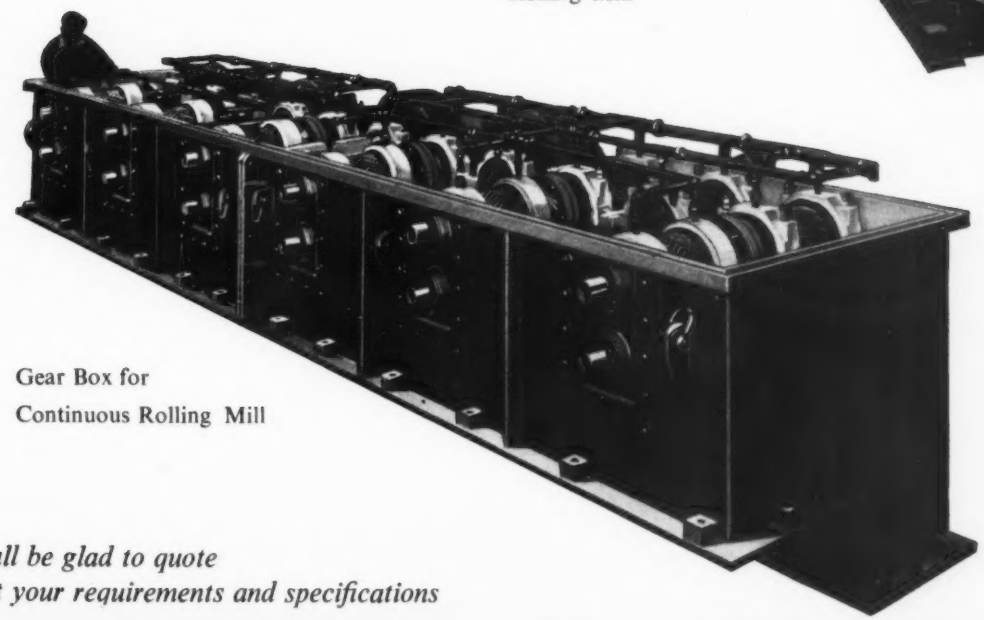

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Problem three was corrosion caused by moisture. Could this be prevented? The answer was given in the Thornton Wet Bearing Rig Test*, where new Shell Alvania won an anti-corrosion rating of 10 out of 10.

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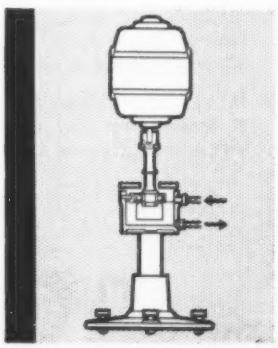
The Research Story

* In the Thornton Wet Bearing Rig Test, water was fed through a 20 mm. bore bearing with a special demountable plastic cage (to simplify inspection) charged with the grease under examination to a weighed quantity rotating at 3,000 r.p.m. The rate of supply of water was one gallon per minute, the water running to waste. After 5 hours' operation the bearing was removed and stored horizontally half immersed in water for three days. Anti-corrosion merit ratings were given at the end of the test. High-quality greases without special protectives showed an average rating of 3 out of 10, while new Shell Alvania rated 10 out of 10.



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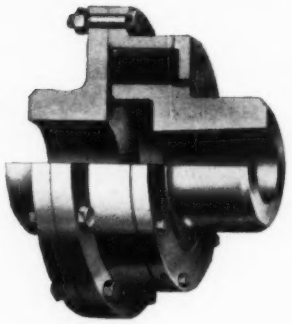

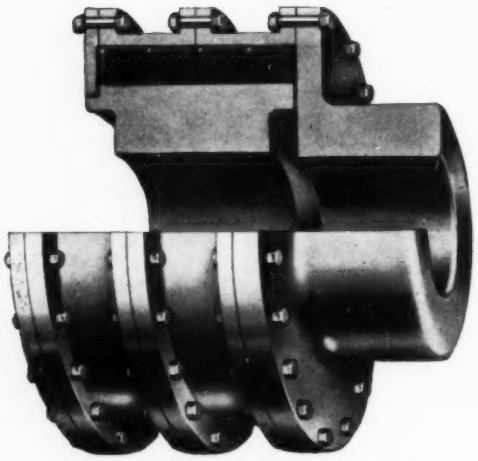

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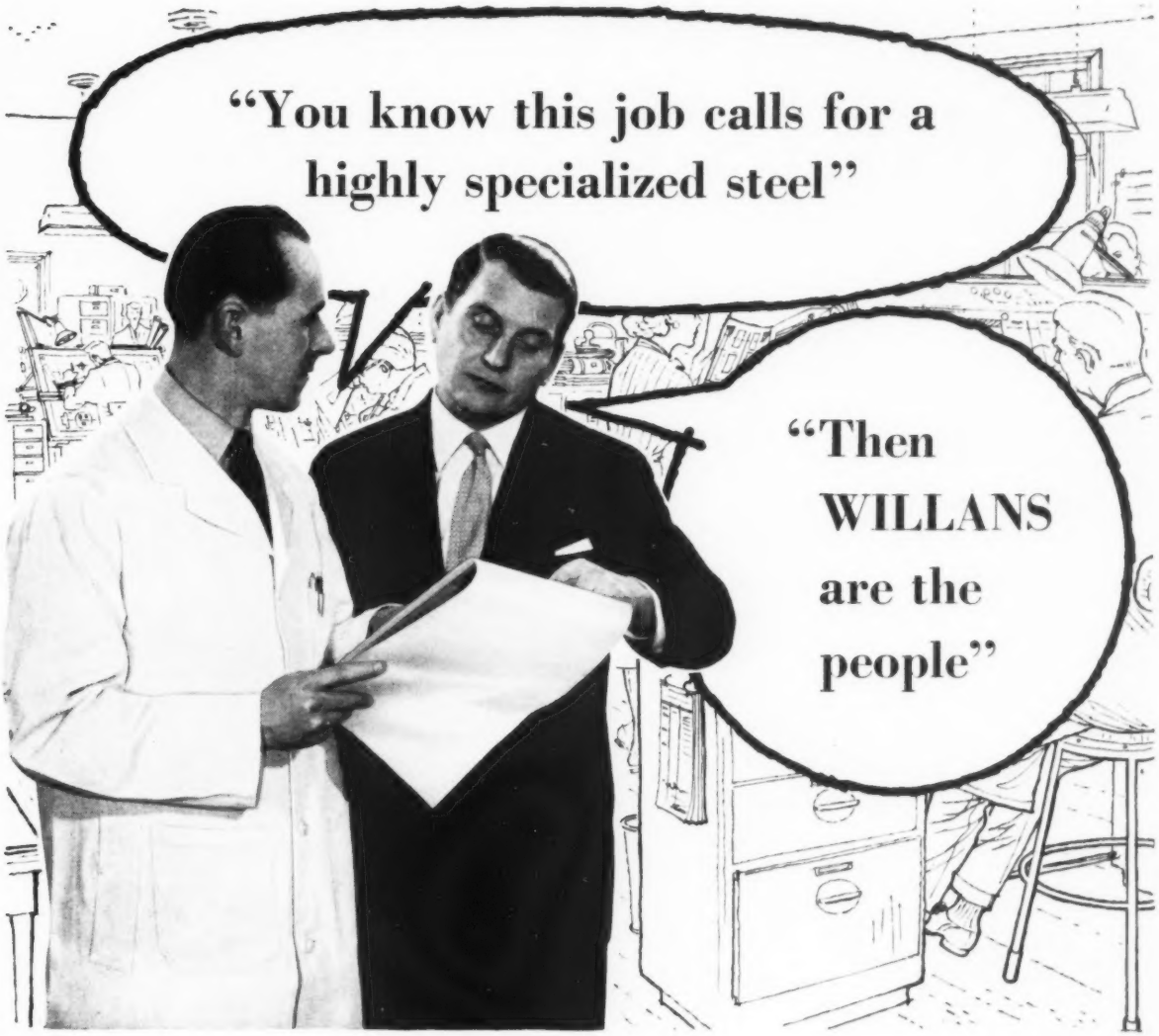

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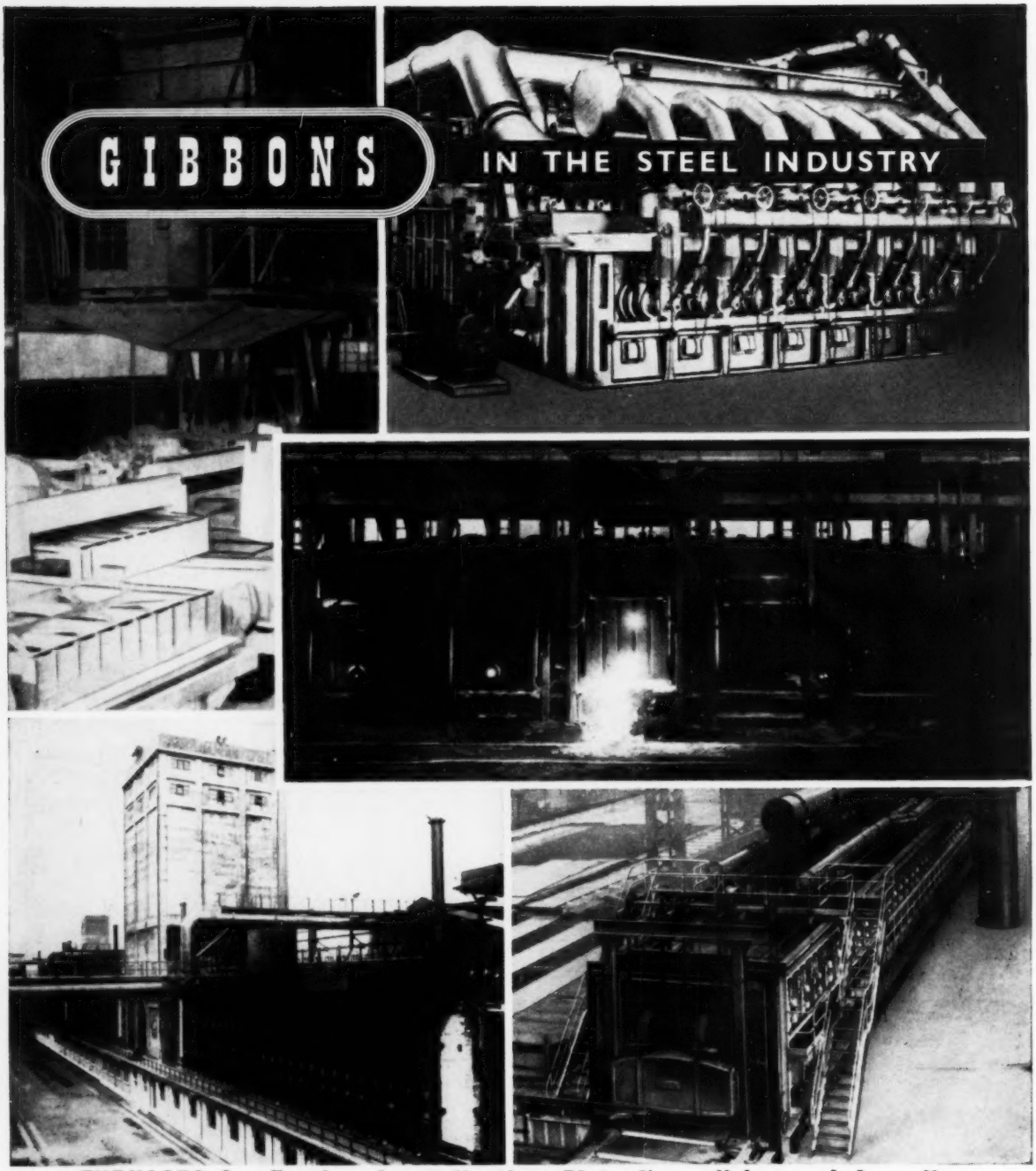

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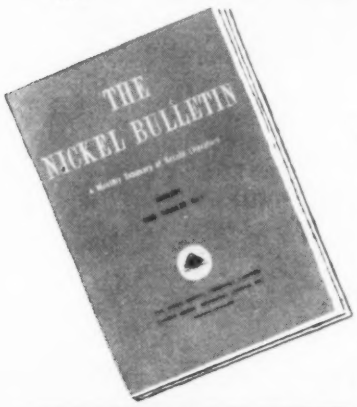

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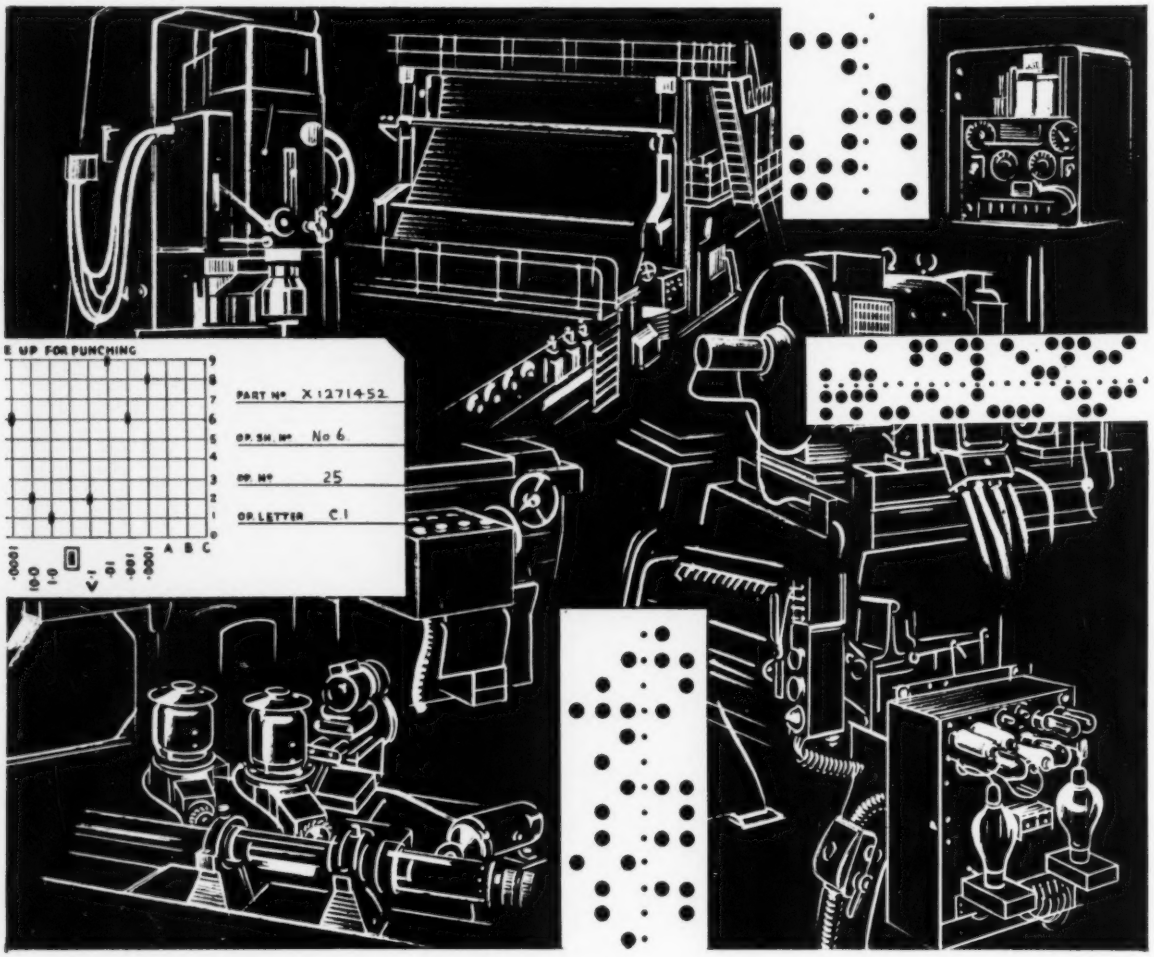

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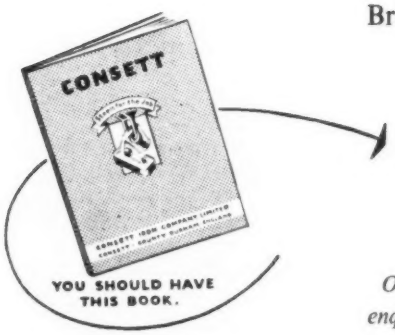

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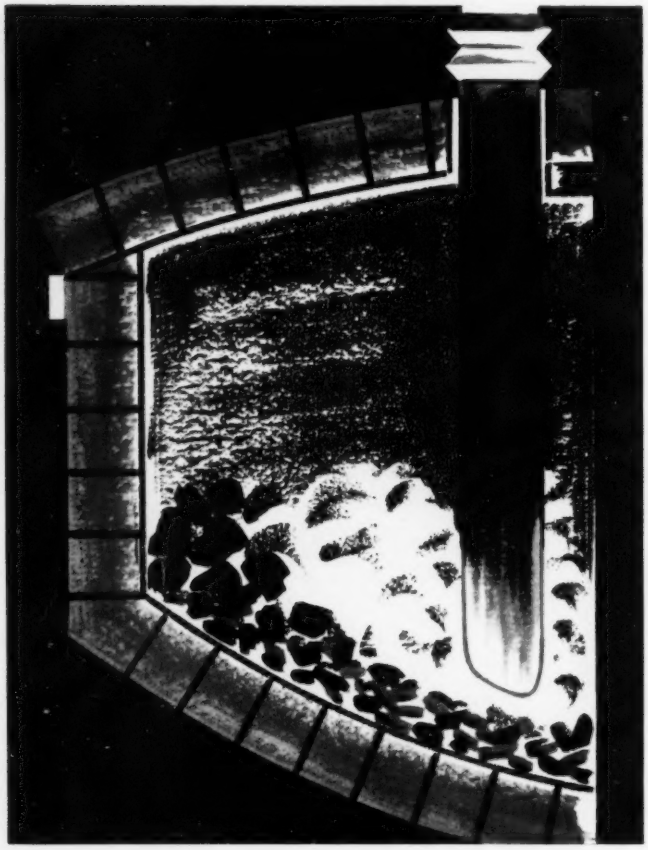

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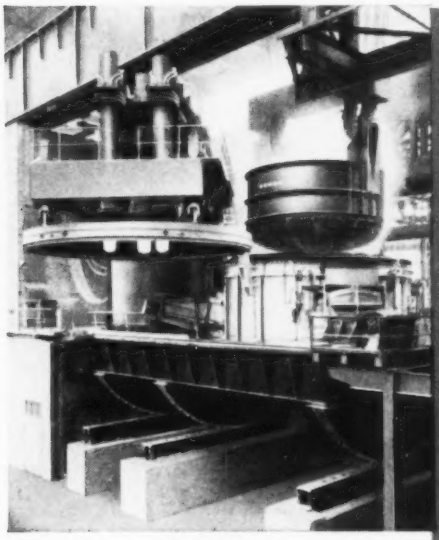

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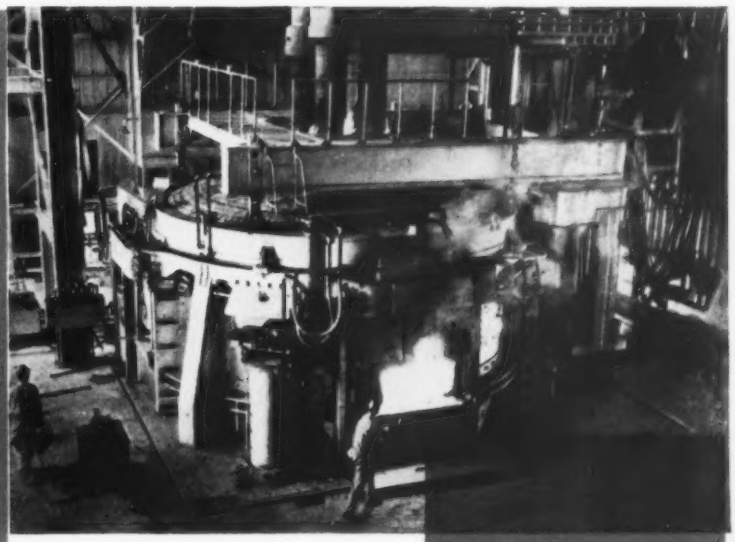

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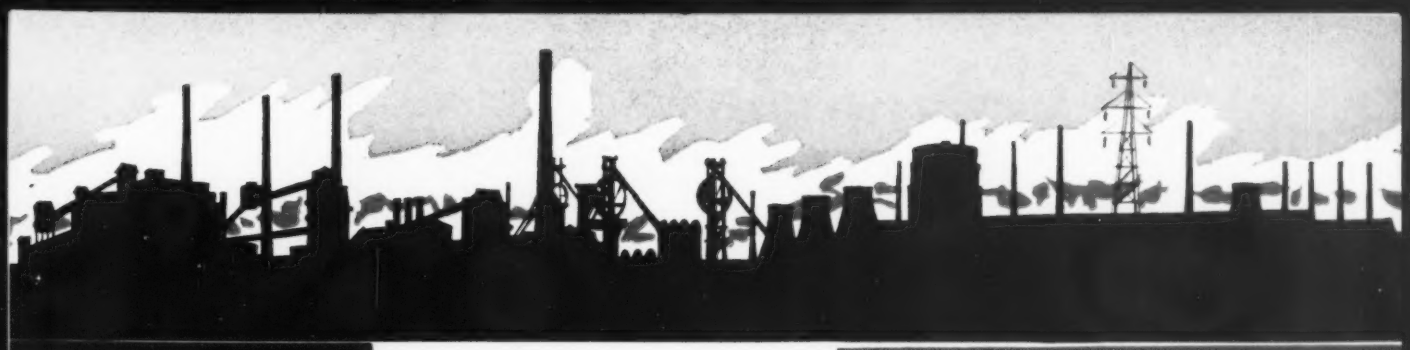

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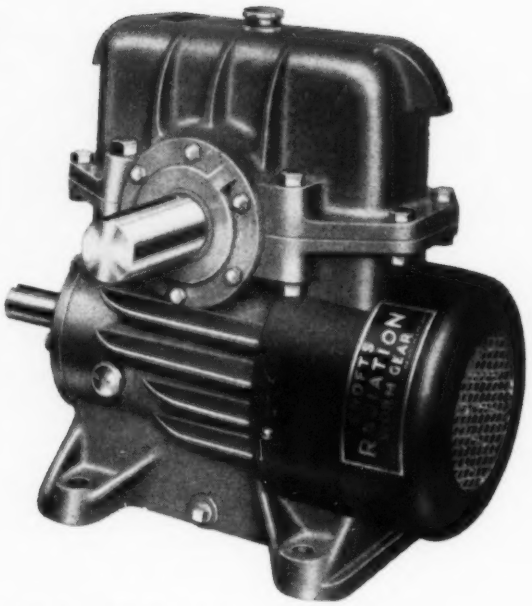

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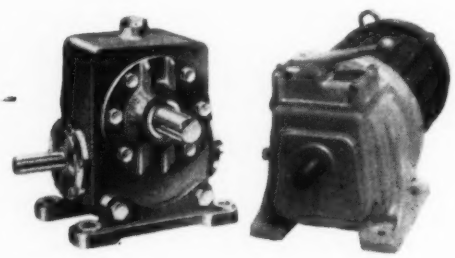

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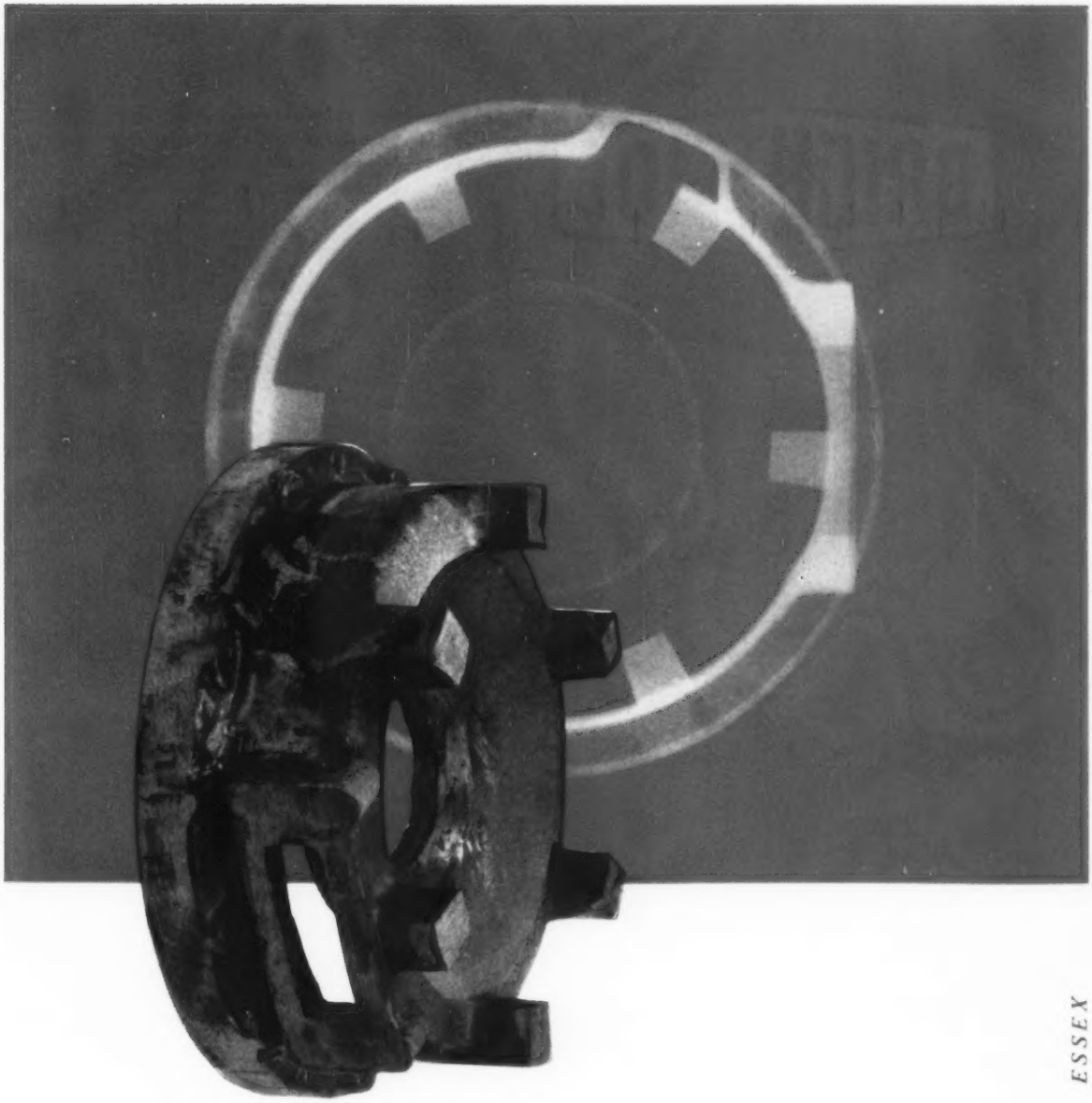

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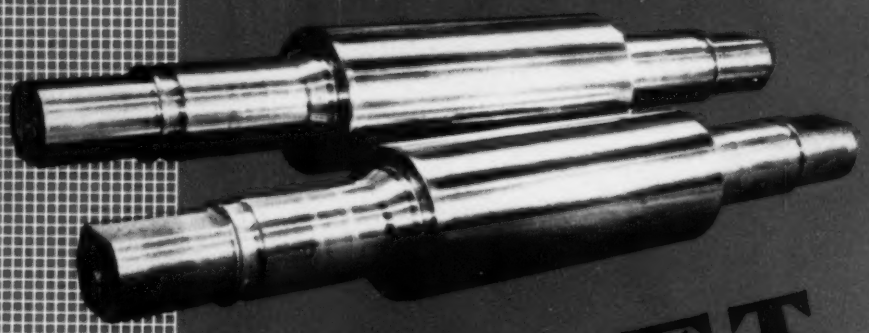

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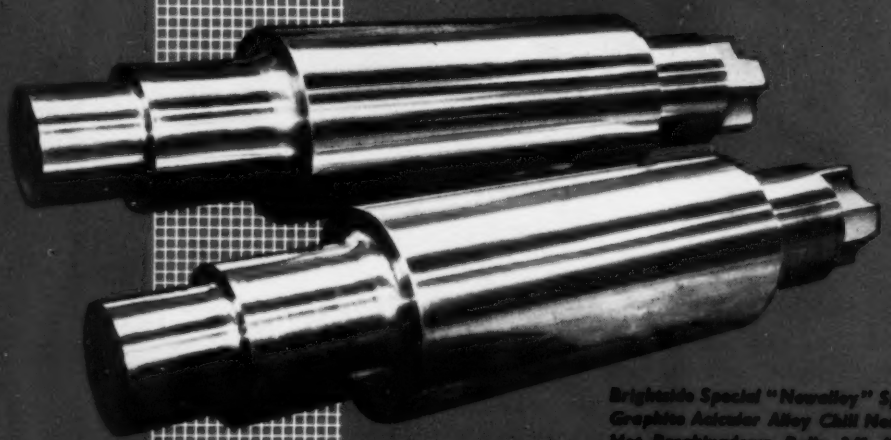

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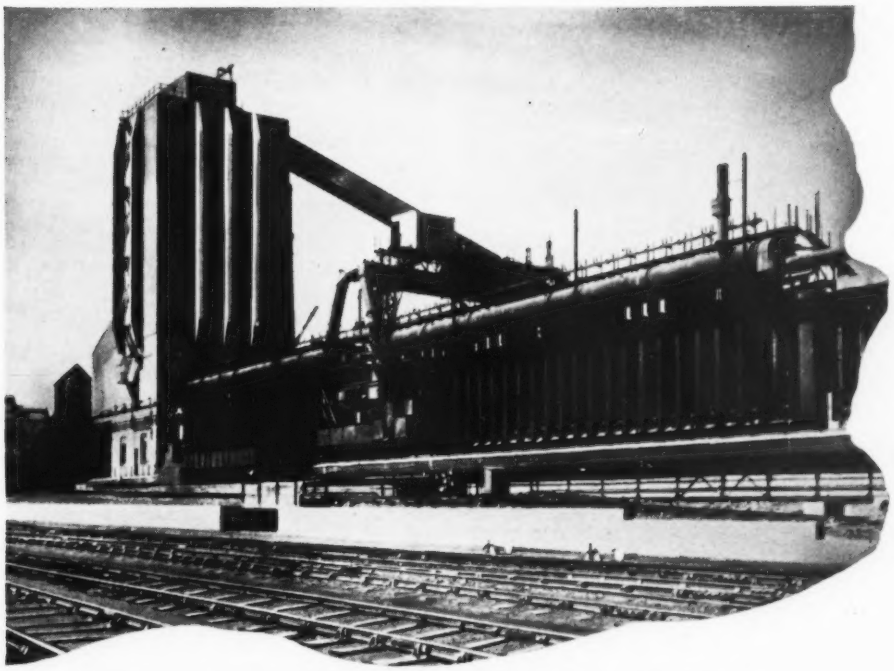

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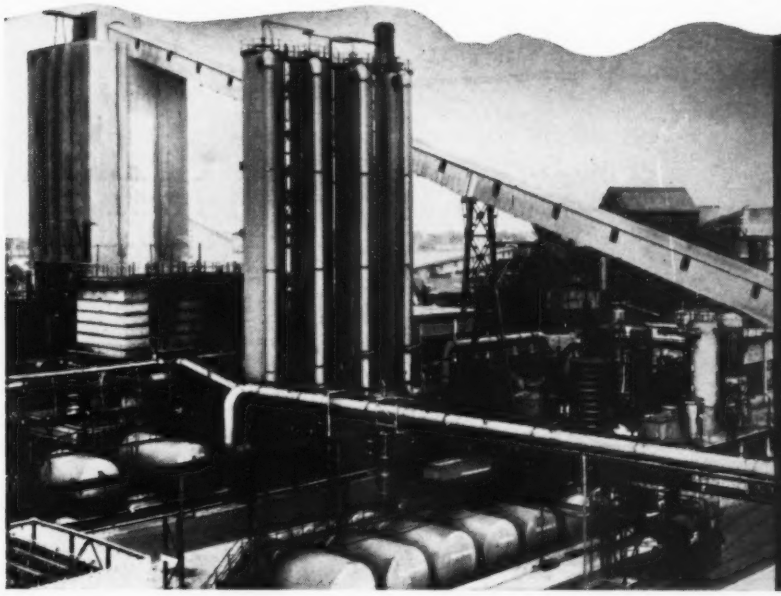

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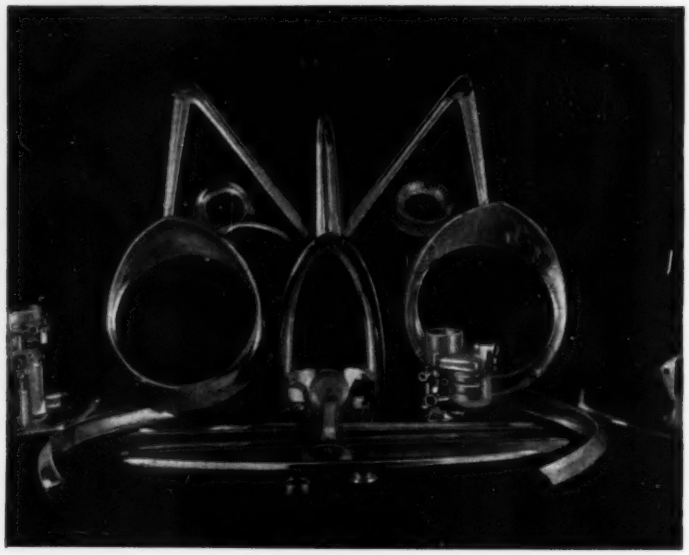

" We decided that Town Gas is the ideal fuel for the Diecasting Process "

Some diecast components made by T.A.L. Developments Ltd., Tottenham.

FUEL SELECTION FOR PRECISION DIECASTING

In August 1945 three men met together and their names were Thomas, Aldridge and Logan. They each put down enough money to create a small capital, and formed a Company which they called T.A.L. Developments Ltd.

Two of them, Aldridge and Logan, having been in diecasting all their lives, felt they knew enough about it to start on their own.

A small factory at Finsbury Park was rented and a labour force of 6 was recruited.

That was the humble origin of T.A.L. Developments Ltd.

Now the firm has a factory of 20,000 square feet, they employ about 120 workers and their capital is still the same. The Company has been built up by ploughing back the profits, until now the firm practically owns a modern factory complete with a modern diecasting plant, with a turnover that has increased nearly threefold in the last five years.

During conversation Messrs. Aldridge and Logan said that prior to 1939 diecasting had not reached the importance in this country that it has now, because large users other than the Motor Industry were the exception rather than the rule.

In 1939 and during the war years, both the Ministry of Aircraft Production and the Ministry of Supply were faced with the necessity of finding a substitute for machined components which were made from the solid bar, and they needed them in large quantities.

Diecasting was their answer. Tools were made for many thousands of intricate parts, and after the usual teething troubles were overcome, these pieces were produced by the hundreds of millions.

The Ministry very quickly realised that a very rigid control of the diecast material was necessary, and so stipulated that all zinc base alloys, used for their parts, were to a rigid British Standard Specification.

This ensured that the diecast parts retained their precise dimensional characteristics which was of vital importance when making delicate fuses, etc.

These facts are well known by T.A.L., who maintain their material to B.S.S., and to prevent any possible contamination use only one zinc alloy, namely, Mazak "3" with a regular analysis taken of material used on every day and night shift. All zinc alloy castings are therefore able to be produced to the British Standard quality, which is a sure sign of top quality.

Mr. Logan, who runs the production side of the factory, forbids the importation of even a small amount of solder into the factory, because of the danger of contaminating the zinc alloy with traces of lead or tin.

There is variety in plenty at the T.A.L. works. Among the larger zinc base pressure diecast components are the motor car headlamp bezels for a popular motor car. These are a complicated design and have to be chromium plated eventually and, to avoid rejects, it is essential that the surface quality of the diecastings is of a very high order. Motor car carburetter parts afford examples of a complex diecasting technique where the diecasting process has minimised machining and drilling costs to a considerable extent. At the other end of the scale there are diecast ladies' watch cases weighing only a few grams and cast so accurately that tiny pieces of marcasite can be fitted into the diecast settings.

Some of the diecasting machines have been bought complete but, what is of extreme interest are those machines which have been built by the firm themselves for their own use to meet special conditions of different contracts.

The diecasting process is well known. The zinc base alloy is melted in a furnace and the molten metal is pumped under high pressure into a steel die which forms the component to the shape of the die. Air valves open and close the die whilst operating the pump in a sequence so that the whole process takes place on a controlled temperature/time schedule. Such machines are built for relatively large, complicated diecastings. On the other hand, when it comes to long runs of relatively small diecastings, a nearer approach to automation is required.

The firm decided to build a number of completely automatic machines. They designed an electronic timer which, by operating the air valves, automatically opened the die, closed it, allowed molten metal to flow into it, then opened it and ejected the casting without any action by the operative, whilst a counting mechanism on the electronic timer told the operative precisely how many components had been turned out.

We turned on to the question of fuel, and on this Mr. Logan was quite emphatic. "Town gas appears to us to be the most suitable fuel for diecasting", he said. "We use gas because of the convenience factor first, then comes its reliability, and the Eastern Gas Board has always been ready to help us out of any troubles which do occur".

Mr. Aldridge added: "Yes, and of course it does mean that we have the simplest form of costing—the meter reading is there for us to see day by day and we can keep a close check upon the fuel cost per unit of output, and of course we know monthly or weekly fuel figures without having to attempt to determine how much solid fuel there is in a heap in the yard or how much liquid fuel there is in a tank".

Mr. Logan took it up from there and said: "You will notice that it is important that all the liquid metal in these diecast-

ing machines is maintained at a controlled temperature. A thermocouple in a protective sheath is located in the chamber containing the molten metal. This thermocouple sends a signal to the control pyrometer which is pre-set to the desired temperature. If the temperature exceeds the set point, then automatically the pyrometer sends an electrical signal to a solenoid valve which in turn operates the valve in the gas supply, so cutting down the gas and allowing the temperature to come back to its correct value. Town gas is therefore an ideal fuel for this process because it has a constant calorific value and is supplied at a governed pressure so that the temperature control is very easily maintained. In addition of course there are no fuel deliveries, no wasted storage space and it is the easiest fuel for the operatives to work with".

T.A.L. began by making zinc base alloy toys, for a large multiple store, and they made many millions of these before gradually changing their output and moving into motor car and general industries. It was interesting to learn that the increasing turnover of the company has not been due to any extensive publicity or to any form of high pressure salesmanship, it being accomplished as a result of doing a job at a reasonable price and producing consistent components so that one satisfied customer mentioned the name of the company to his friends in associated industries. That, with Mr. Aldridge's extensive technical knowledge, enabling him to give quick and accurate competitive costing is the true story of the growth of the company.

It is only when one visits a diecasting firm such as T.A.L. Developments that one realises the extent to which zinc base alloy diecastings fit into our pattern of life. There are gold plated zinc base diecastings for electric shavers. There are components for electrical radio and television sets made for leading firms in the electrical and radio industries—washing machine components—carburetter parts, etc.

Not only do the firm specialise in zinc base diecastings, but they also make considerable amounts of pressure aluminium alloy castings. The process here is somewhat different. The aluminium alloy is separately heated by gas in non-metallic crucibles, again on pyrometer control principle, and is ladled in the molten state into the diecasting machine. It is thus a cold chamber process as compared with making zinc base diecastings, which of course is a hot chamber process because the pressure cylinder is immersed in the molten metal.

Ever since 1945 the local Gas Board has collaborated with T.A.L. Developments —its Industrial Officers helped in the provision of piping when supplies were difficult, they made recommendations and acted as Consulting Engineers when the roof of the new factory had to be insulated to prevent heat wastage and the firm required a loan from the Ministry of Fuel and Power for this work to be done. The result of this co-operation is shown in the steady increase in the industrial gas sales load from the Eastern Gas Board to the company, which has doubled itself in the last five years.

"Some say town gas is an expensive fuel", was a remark put to him, and he smiled and said: "Well, I suppose any raw material used to make an end product is considered to be expensive by the manufacturer".

"But it is all a question of relative terms. We have a free choice to use any fuel we would wish and we have always used gas and see no reason why we are likely to change in the future. It is clean, it is provided at a constant calorific value. It is a flexible fuel which lends itself very easily to automatic temperature control so essential in our process, there is no storage space, there are no ashes to clear away and in short it is a piped fuel coming direct to us from the gas works. When we compared all these advantages and carefully worked out the relative fuel costs we decided that town gas was the ideal fuel for the diecasting process".

Each gas user has the advantage of the gas industry's pooled knowledge of so many different industries. An Industrial Gas Engineer of the Eastern Gas Board, for example, is always able to make use of the knowledge of the Industrial Officers in all the other area Gas Boards and has access to the Gas Technical Bureau in London, where a close watch is kept upon all progress in gas-fired equipment not only in the United Kingdom but also in overseas countries.

Precision diecasting is just one of the some 4,000 trades, professions and occupations benefiting from the constant research and development taking place in the Gas Industry.

"Yes—it does mean that we have the simplest form of costing".

Typical diecasting machine open, showing the die. Molten metal heated by town gas is forced into the die.

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Northern Gas Board, Newcastle-upon-Tyne.
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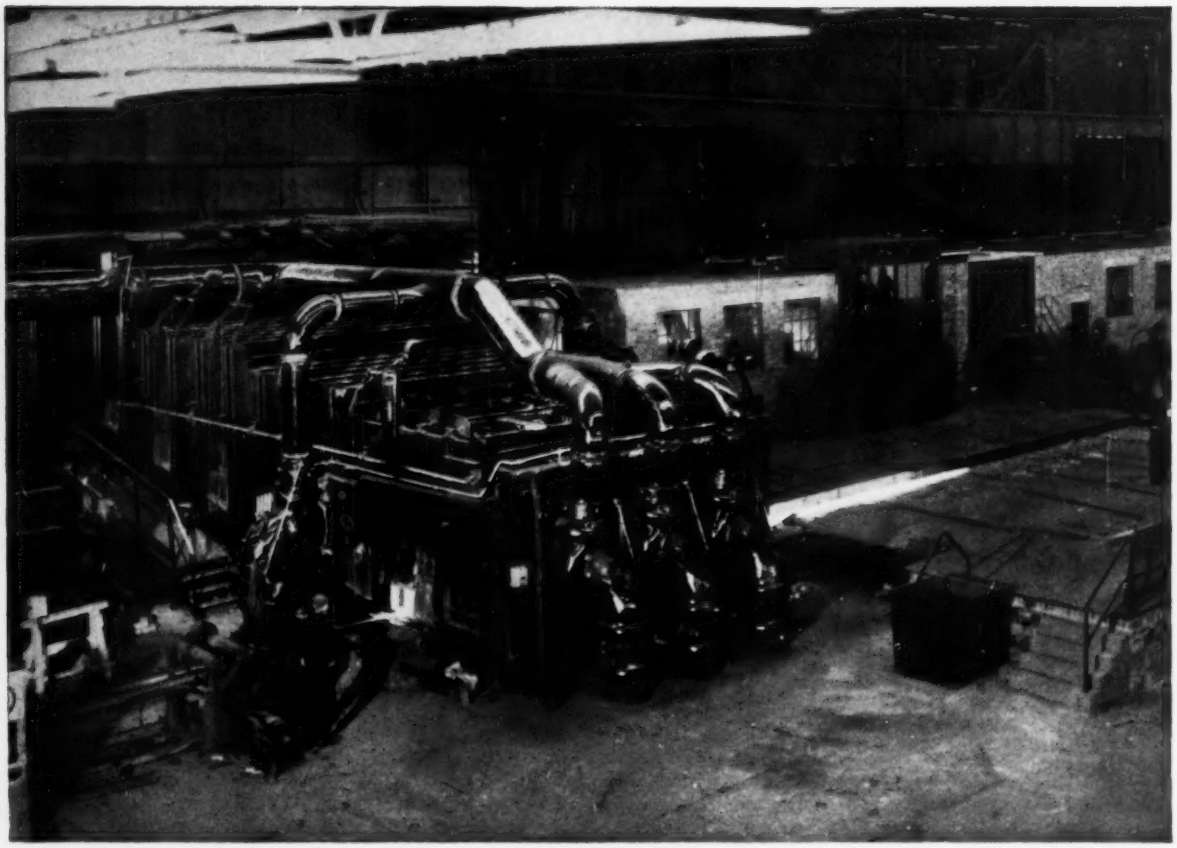

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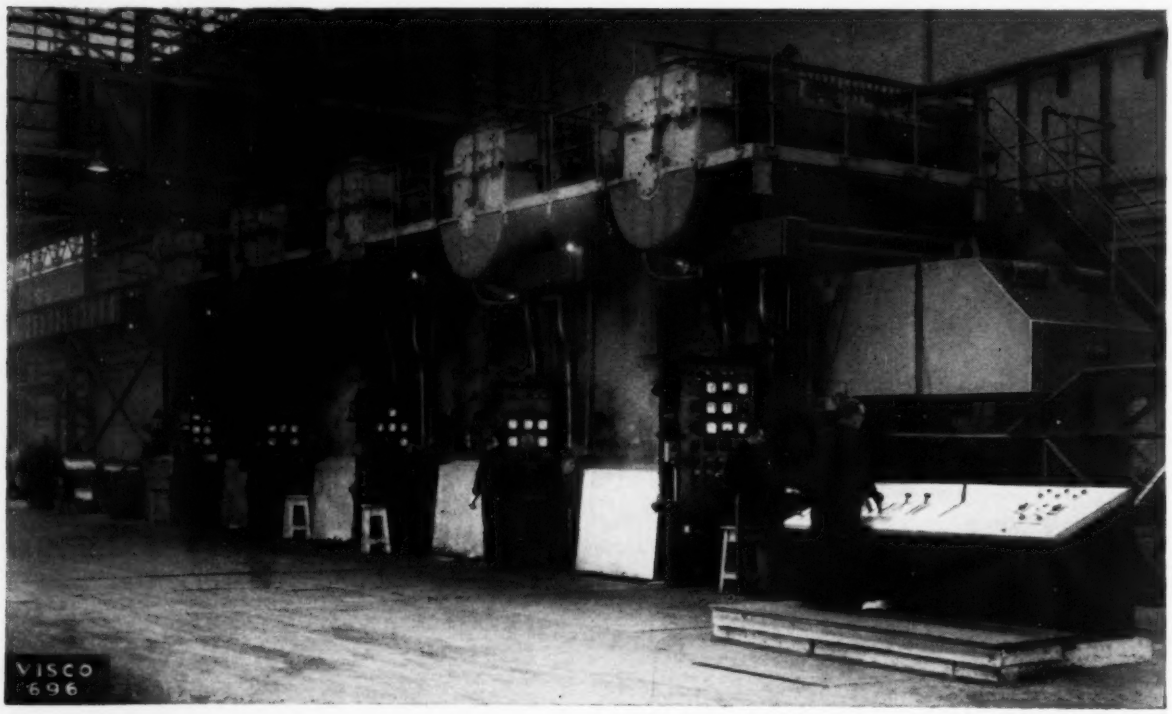

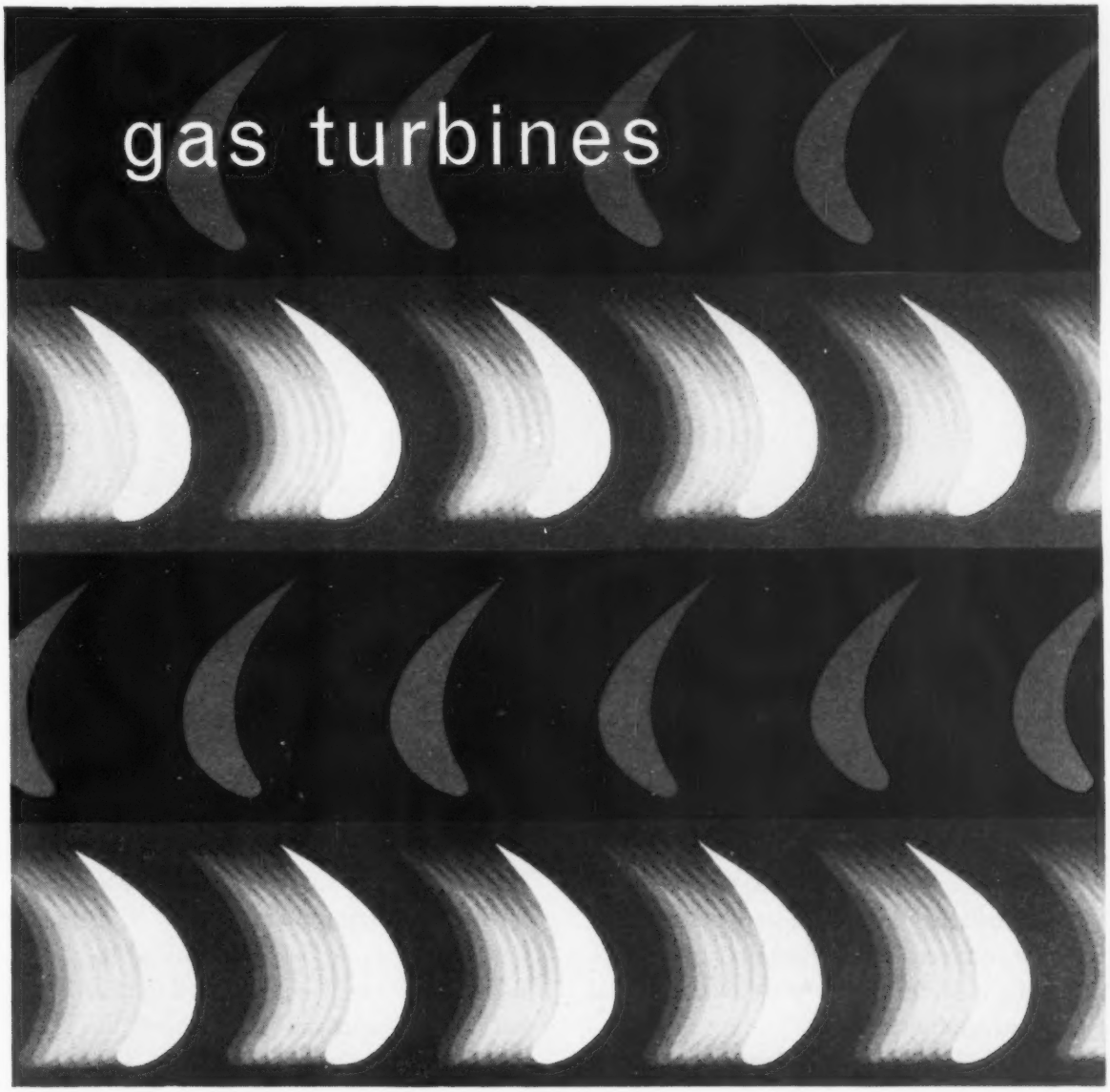

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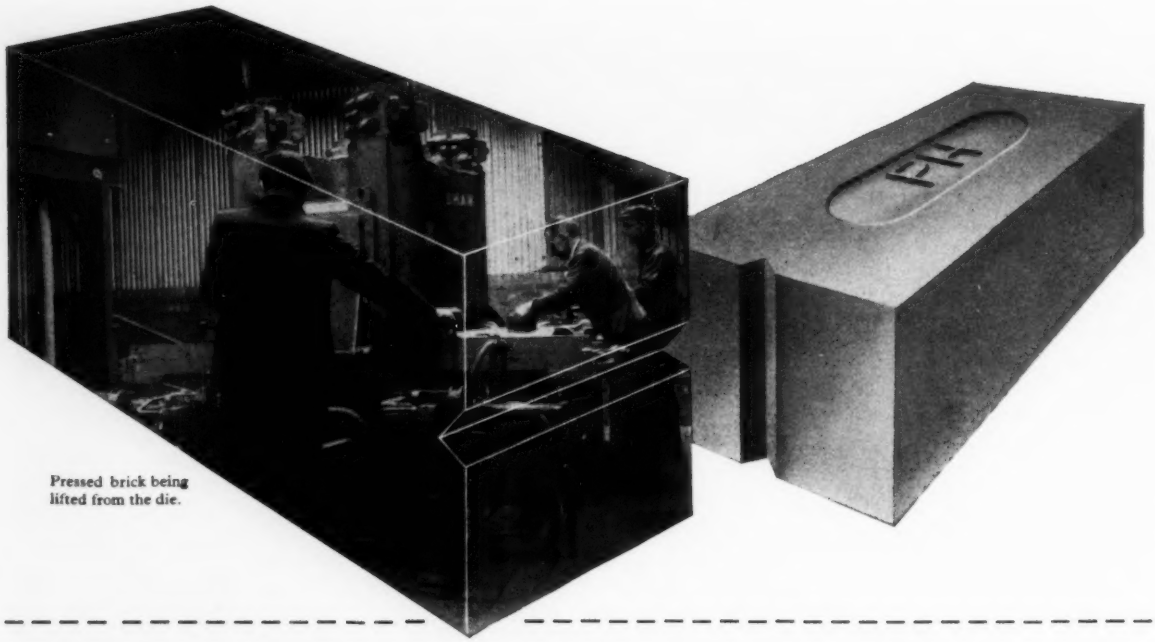

Pressed brick being lifted from the die.

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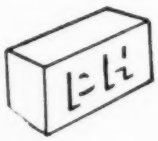

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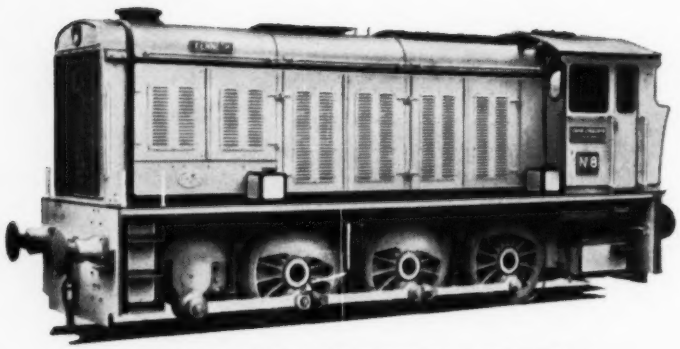

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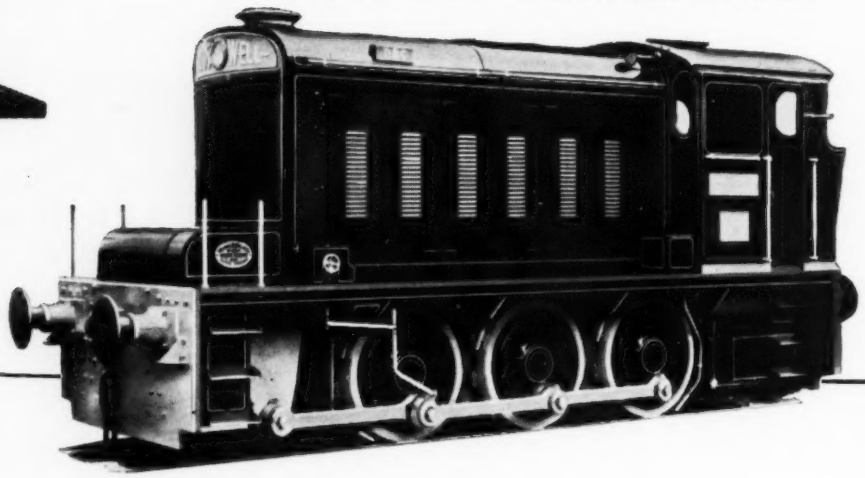

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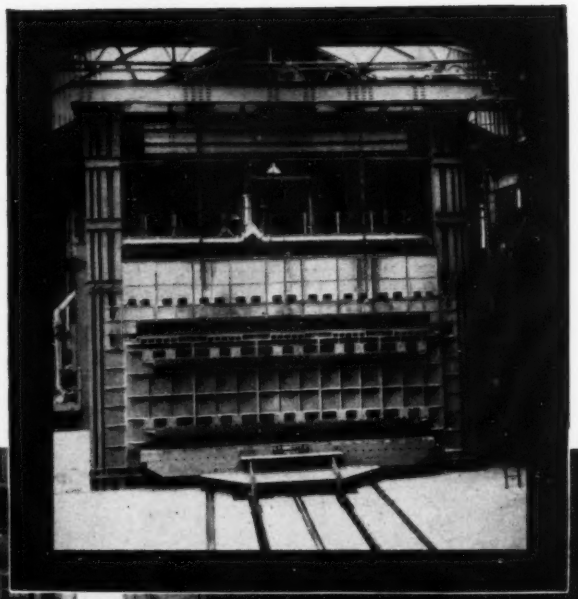

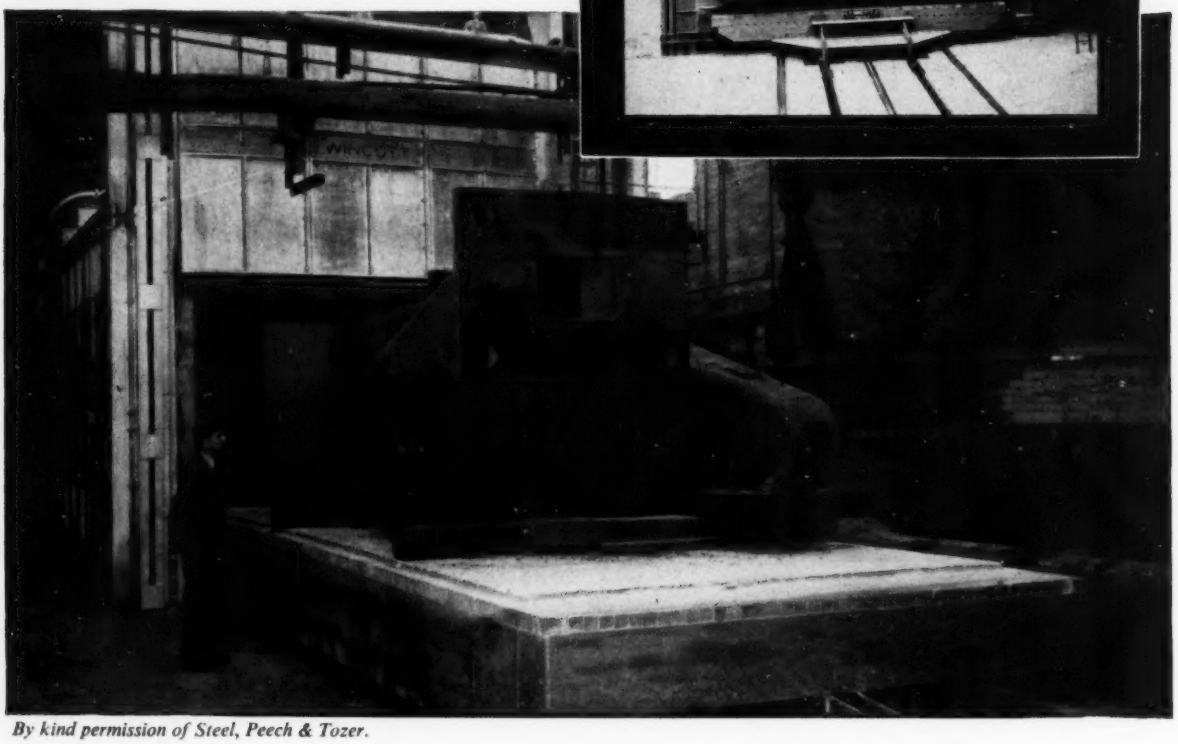

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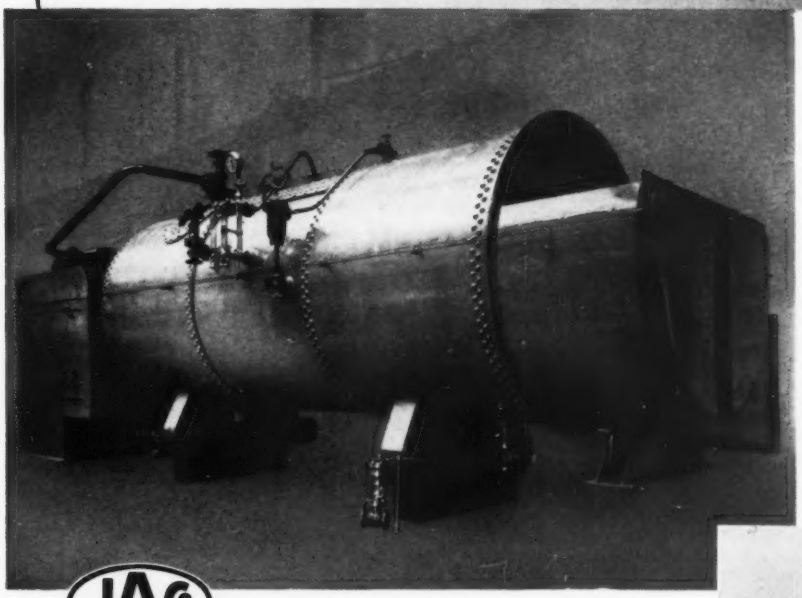

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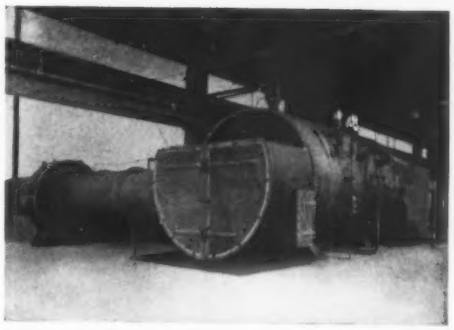

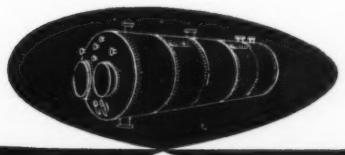

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So famous have these amazing drills become that the name DURIUM tends to be used for any drill of its general type. *Don't be misled by this.* DURIUM means the special New Process, almost diamond-hard, carbide tip developed and made only by the Rawlplug Company. *No* other drill is a genuine Durium. DURIUM Drills give really sensational cutting performances in brick, tile, cement, slate, etc. They are silent in operation and have a quite exceptional long-service capacity for withstanding the abrasive action of masonry that so quickly ruins ordinary drills. Another exclusive feature of Durium Drills is their patented rapid-helix flute, which forces out spoil and prevents clogging. Only a Durium Drill—a *genuine* Rawlplug Durium—gives such life and service. First resharpening is FREE!

For use in Wheelbrace or slow speed electric drill

17 sizes 5/32" to 1" cutting diameter and **11** *long series* ¼" to 1" cutting diameter for up to 16" drilling length.

'DURIUM' TIPPED HAMMER DRILLS

For percussion drilling in hard materials such as Granite, Blue Bricks, Concrete etc., Durium Hammer Drills have a very much longer life than ordinary percussion drills. They can be used in most electric Hammers and some type of Pneumatic Hammers. 13 sizes from 7/16" nominal drill diameter to 1¼".

'DURIUM' TIPPED GLASS DRILLS

For use in glass, china, vitrolite, pottery, etc. Can be used in a wheelbrace or slow speed electric drill.
9 sizes from ⅛" to ½" cutting diameter.

'VIBROTO'
HARD TIPPED DRILLS FOR VIBRATORY DRILLING

Vibroto Drills are specially manufactured for use with the Vibroto Drilling Machine, —a high efficiency power tool which allows of two vibratory actions (Light Rapid and Heavy Slow), and also a rotary action for use with standard Durium masonry Drills or ordinary Twist Drills. With their special cutting angle metallic carbide tips, and the high grade steel shanks heat-treated to withstand abrasion, Vibroto Drills put up spectacular performances in the drilling of concrete and other hard materials. *Demonstrations* can be carried out by Rawlplug Technical Representatives anywhere in the British Isles.
10 sizes 5/32" up to 15/32" cutting diameter.

'RAWLCRETE'
TUNGSTEN CARBIDE TIPPED (ROTARY) CONCRETE DRILLS

These drill are designed for hole boring in concrete, hard brickwork and tiles with portable rotary action tools. They have from four to six cutting edges (depending on size) and if one should be damaged the remainder will go on cutting. Spoil is automatically carried away through the cored shank and ejected via the slot.

EXTENSION TYPE DRILLS ANY DEPTH

By adding extension rods to the Extension Type Rawlcrete Drill, any depth of hole can be drilled.
26 sizes ¼" up to 2" cutting diameter.

RAWLPLUG FIXING DEVICES AND TOOLS FOR SPEED AND RELIABILITY

WRITE NOW
for illustrated literature describing fully these and other Rawlplug high performance drills and power tools.

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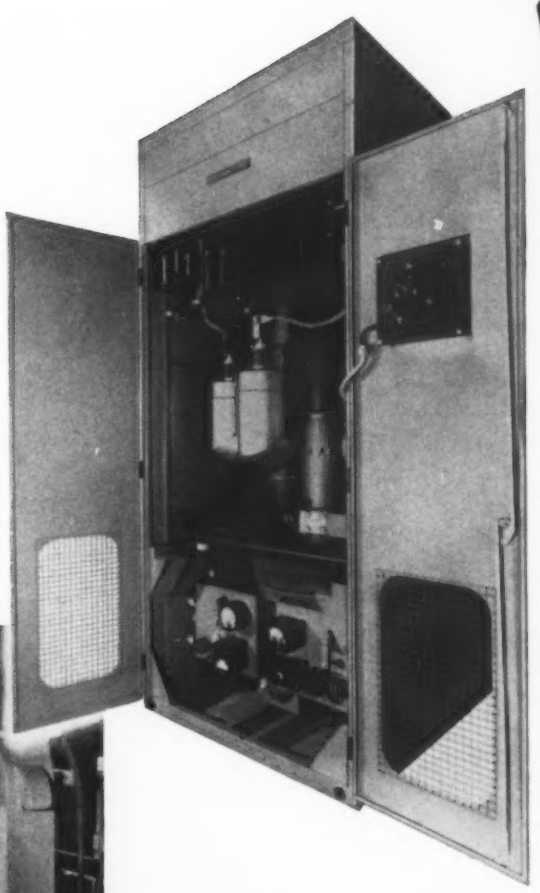
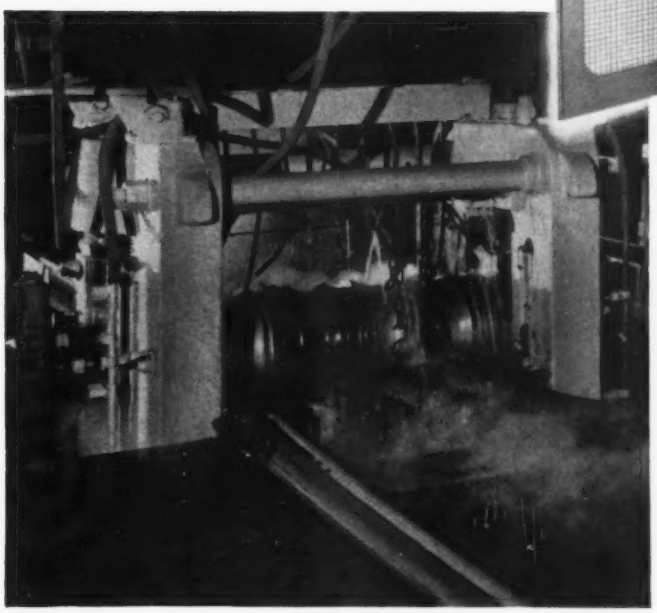
The World's Largest Manufacturers of Fixing Devices.

Britain's first

converter-fed hot reversing mill





The first two hot reversing mills in Britain to be fed from mercury-arc converters have been engineered and commissioned by 'ENGLISH ELECTRIC' at the Darlington steelworks of Messrs. Darlington and Simpson Rolling Mills Ltd.

This technique greatly reduces the number of rotating machines and their heavy foundations, resulting in lower initial and running costs, with minimum maintenance.

Over 50 years world-wide experience stands behind 'ENGLISH ELECTRIC' service to the metal industries and is available to help in the solution of your particular problems.

'ENGLISH ELECTRIC'

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◀ *80 ton steelworks crane.*

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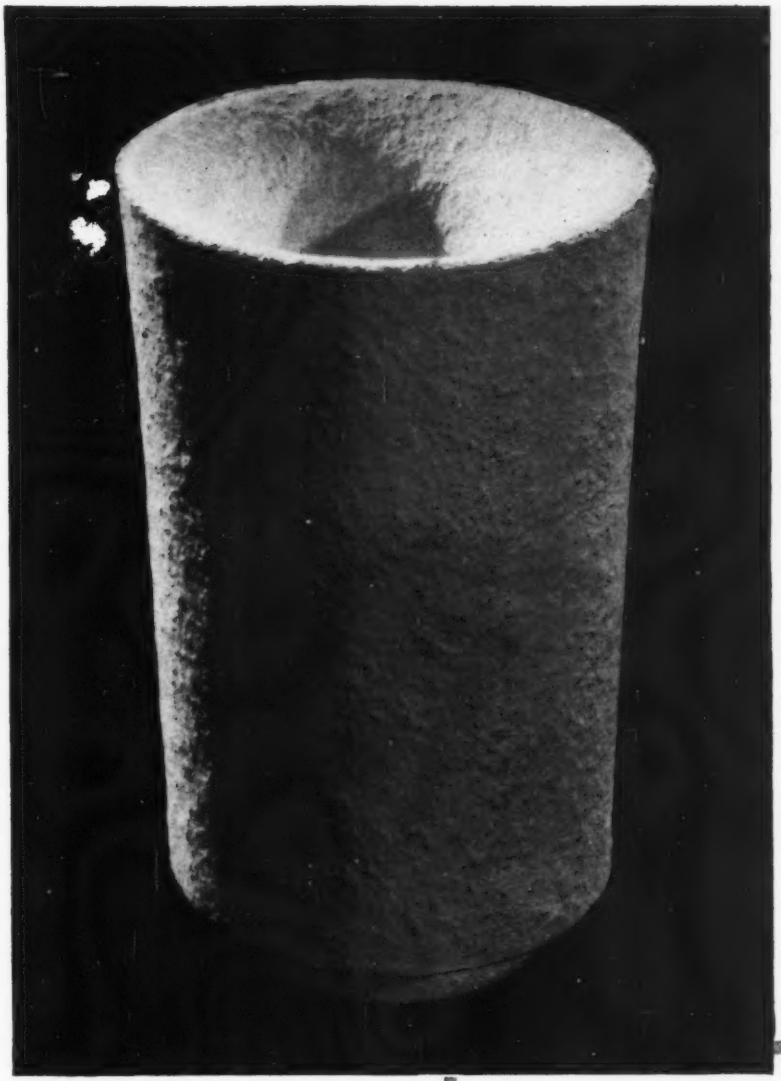

Nozzle nonpareil

Wraggs of Sheffield were pioneers in the manufacture of casting pit refractories, and have always concentrated their efforts in this field. The results of this specialisation can readily be seen in, for example, this nozzle—in which only the finest materials have been used for reliability and general all-round efficiency.

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STOPPERS

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TRUMPET BELLS

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Coils or cut lengths

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Speciality:

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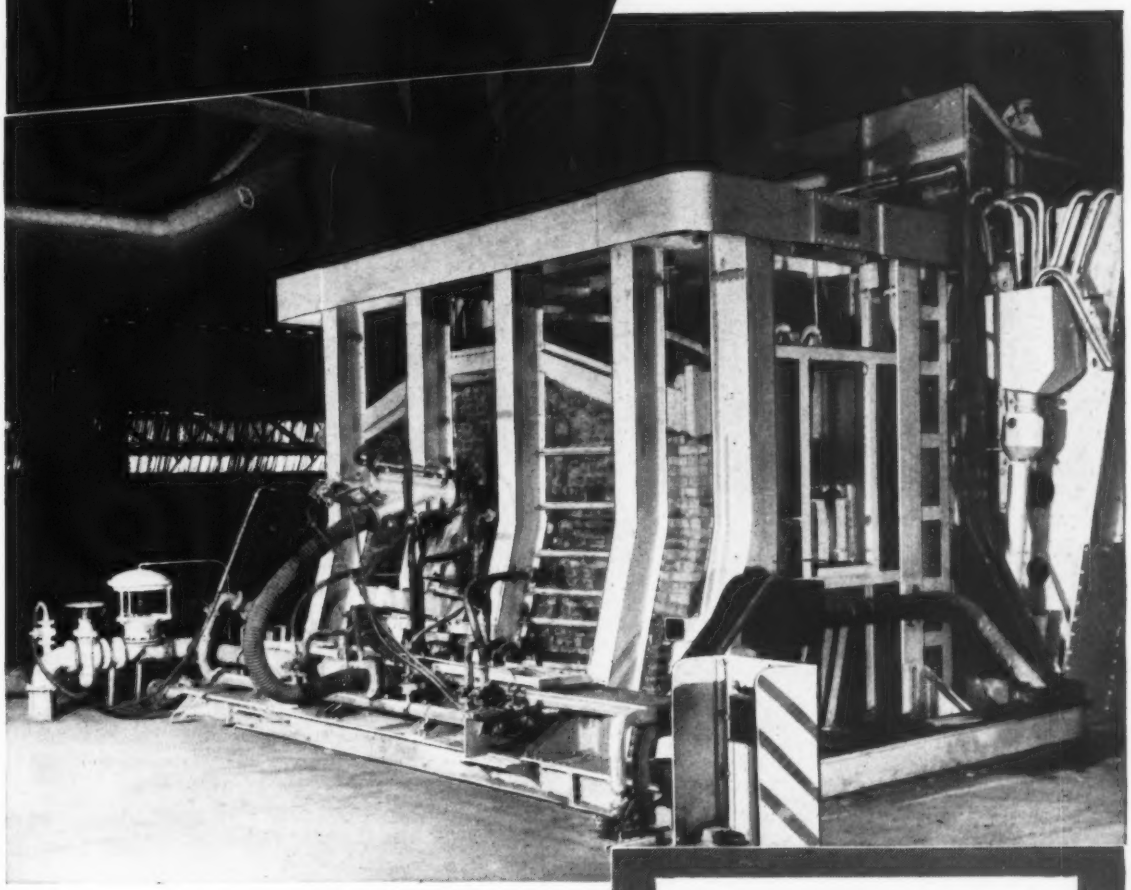

The above shows the application of a Maerz Port End to a 350 ton Tilting Furnace, resulting in greatly improved performance and reduced maintenance costs. This is one of a number of similar conversions carried out at the Clydebridge Works of Colvilles Limited.

Priest Furnaces Ltd.

Sole representatives in Britain for **MAERZ** *open hearth furnaces*

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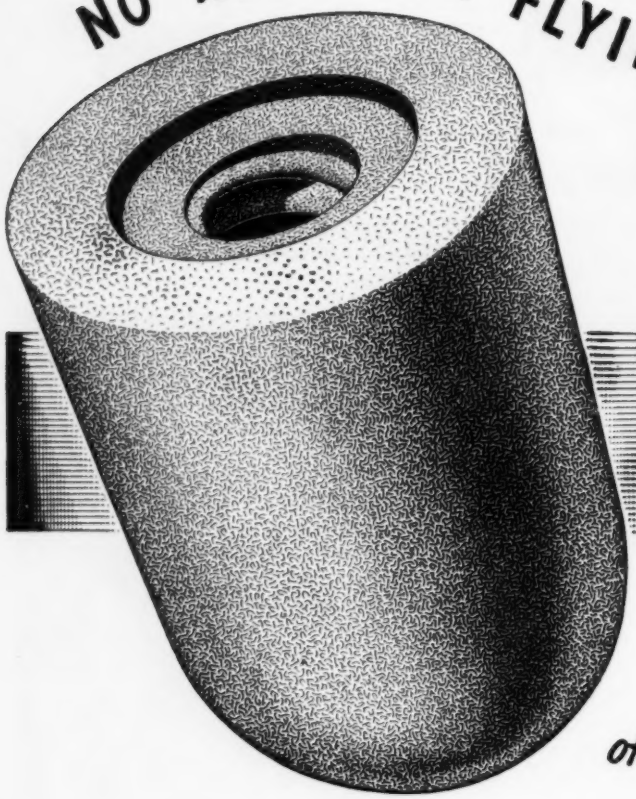

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SR

SPALL RESISTANT

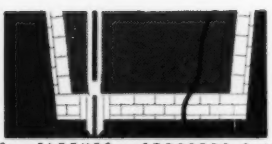
STOPPER END

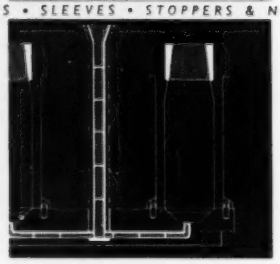
offers EXTRA *Reliability*

The focal point of teeming. The Stopper End. Subject to thermal shock, intense wear by liquid steel and severe mechanical stresses at high temperatures. No half measures with stoppers—they work or fail.

Resulting from years of development work Marshall's now present their "SR" quality. Already acclaimed and well-known in the biggest ladles of the major integrated plants, "SR" is equally satisfactory in the steel foundry and when teeming the smaller 'special' steel charges.

"SR" is highly refractory, accurately shaped—and tough. Consistently better and costs no more.





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One of the MARSHALL REFRACTORIES GROUP of Companies

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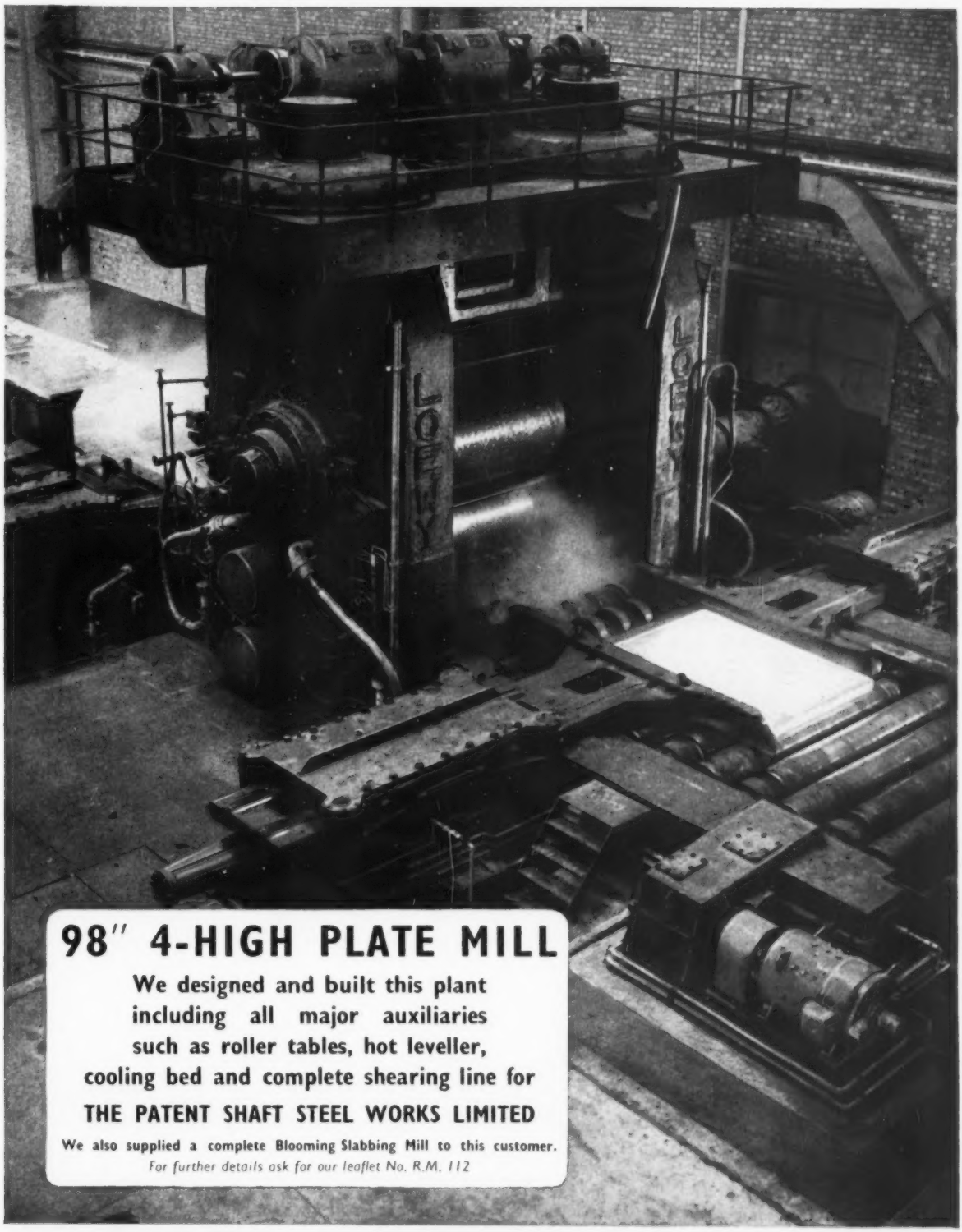

98″ 4-HIGH PLATE MILL

We designed and built this plant including all major auxiliaries such as roller tables, hot leveller, cooling bed and complete shearing line for

THE PATENT SHAFT STEEL WORKS LIMITED

We also supplied a complete Blooming Slabbing Mill to this customer.

For further details ask for our leaflet No. R.M. 112

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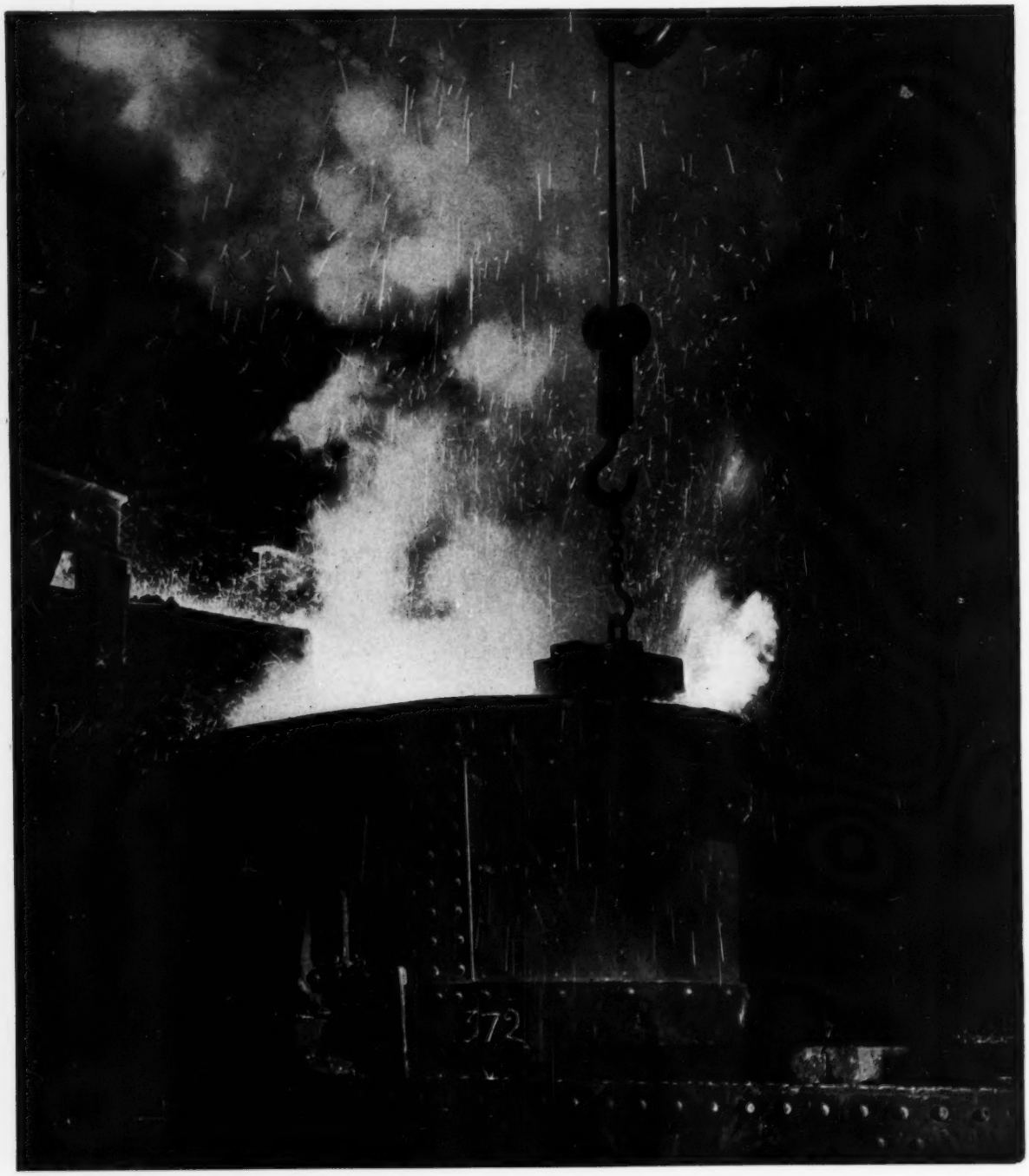
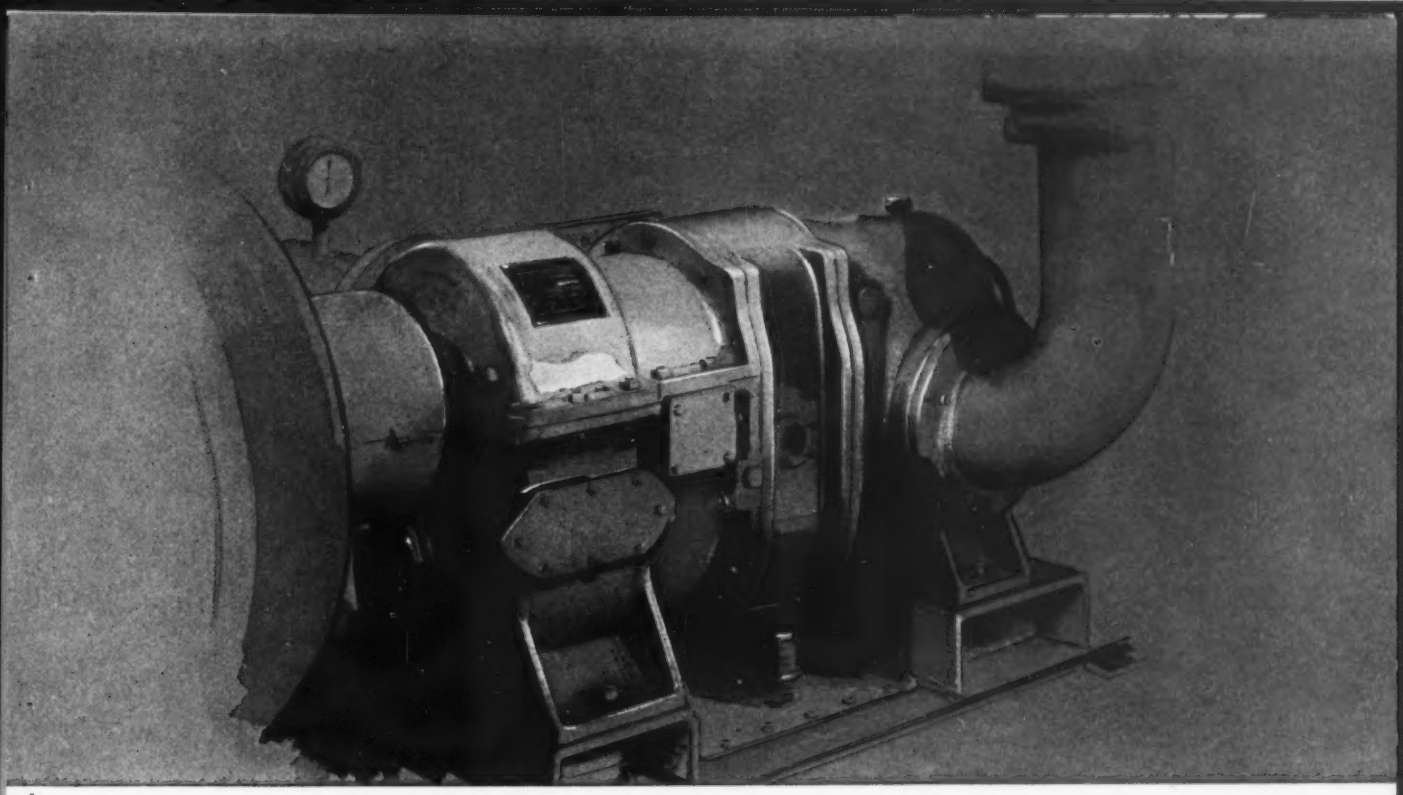
BOURNEMOUTH ENGLAND



BROWN BAYLEY STEELS LIMITED SHEFFIELD

BROWN BAYLEYS



Howden rotary-screw compressor, 1100 c.f.m., exhausting at 1/5 of an atmosphere at the AIRCRAFT RESEARCH ASSOCIATION, BEDFORD

Howden Compressors

For more than 12 years, Howden have been manufacturing and developing rotary-screw compressors. Howden compressors have many outstanding advantages, as follows:—

As rotary-screw compressors without any reciprocating parts they are fundamentally free from out-of-balance forces, with a large output in relation to their size. With their high speed and freedom from valves they give a pulseless delivery.

Accurate gearing maintains the rotors with a designed clearance; thus they never touch and are less affected by air-borne impurities. The rotors need no lubricant; the air or gas delivered is therefore oil-free, an advantage in certain industries and an economy in all.

Howden compressors are driven through Howden speed-increasing gearboxes from any suitable power source.

ECONOMICS OF THE COMPRESSOR

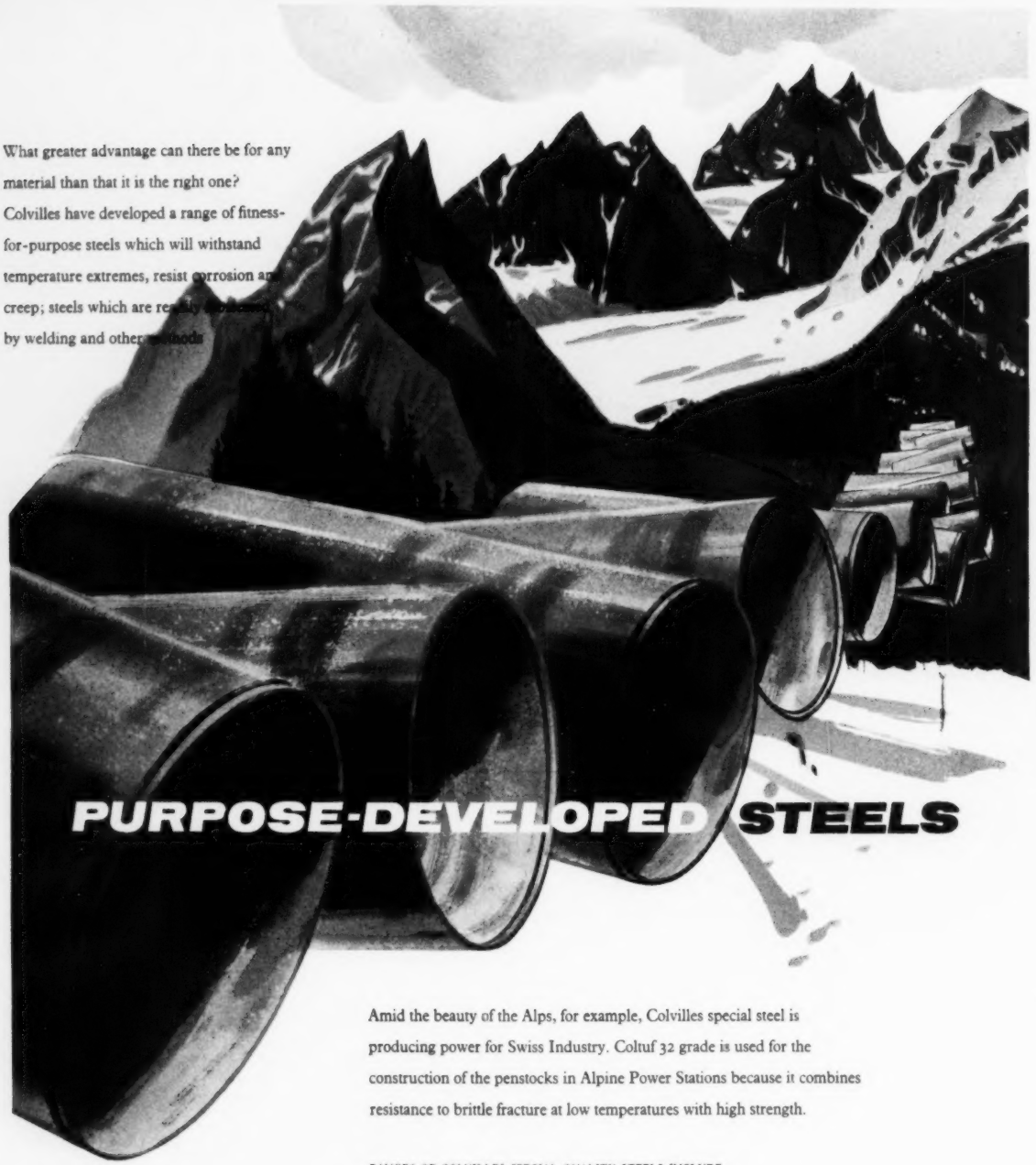
The simple mechanical construction reduces maintenance costs and increases availability. Cost of foundations is lower. Floor space occupied is smaller and receiver needed is smaller.

JAMES HOWDEN & COMPANY LIMITED

195 SCOTLAND STREET, GLASGOW, C.5 and 15 GROSVENOR PLACE, LONDON, S.W.1

HOWDEN

What greater advantage can there be for any material than that it is the right one? Colvilles have developed a range of fitness-for-purpose steels which will withstand temperature extremes, resist corrosion and creep; steels which are readily fabricated by welding and other methods



PURPOSE-DEVELOPED STEELS

Amid the beauty of the Alps, for example, Colvilles special steel is producing power for Swiss Industry. Coltuf 32 grade is used for the construction of the penstocks in Alpine Power Stations because it combines resistance to brittle fracture at low temperatures with high strength.

RANGES OF COLVILLES SPECIAL QUALITY STEELS INCLUDE:

ABRAZO Wear-resisting steel
COLMO Creep-resisting steels
COLTUF Notch Ductile steels
COLCLAD Stainless Nickel and Monel clad steels
CORTEN Corrosion-resisting steel
DUCOL W.30 Weldable high strength steel

ENQUIRIES ARE INVITED FOR THESE AND OTHER SPECIAL QUALITY STEELS

COLVILLES LIMITED STEELMAKERS TO THE WORLD

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MAKING LARGE TIMKEN BEARINGS

FOR 30 years British Timken have been making large Timken tapered-roller bearings.

To meet the ever-growing demand it was decided five years ago to build a new factory at Daventry for manufacturing large bearings only, and this factory has now been in full operation for more than two years.

Some Timken bearings exceed 7 feet in diameter, others weigh as much as 4½ tons.

In this view of part of the Machine Shop, high-nickel-alloy steel forgings are being machined prior to heat treatment.

TIMKEN REGD. TRADE MARK

TAPERED-ROLLER BEARINGS MADE IN ENGLAND BY

BRITISH TIMKEN

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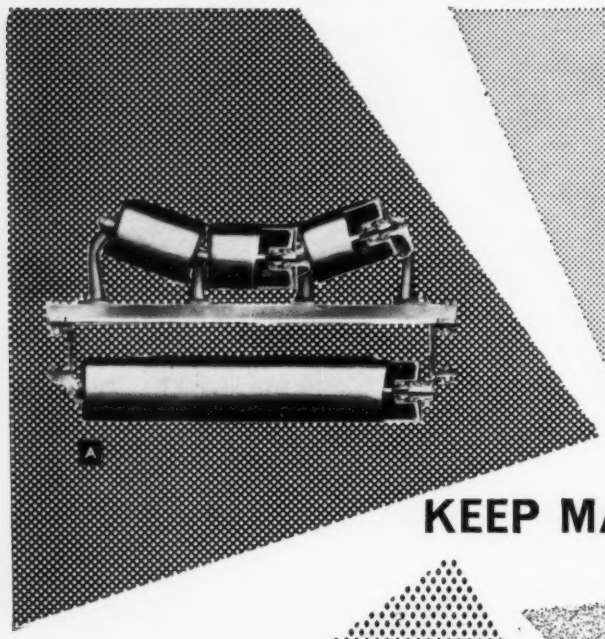
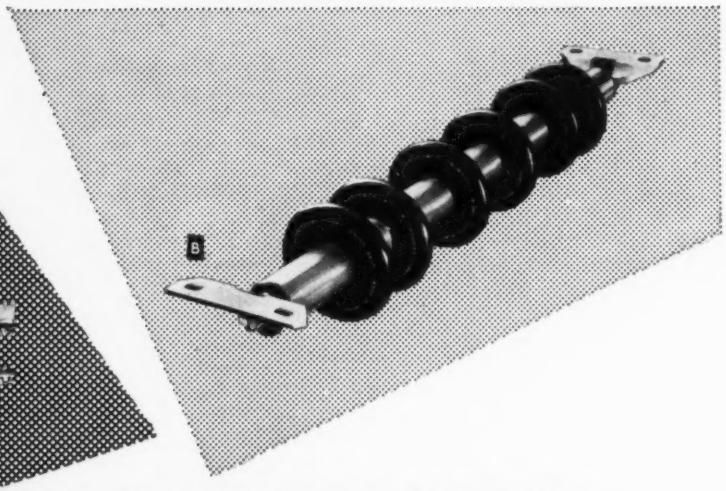
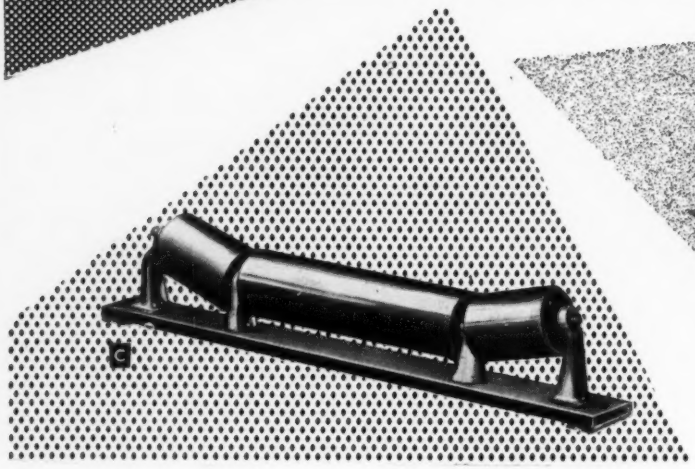
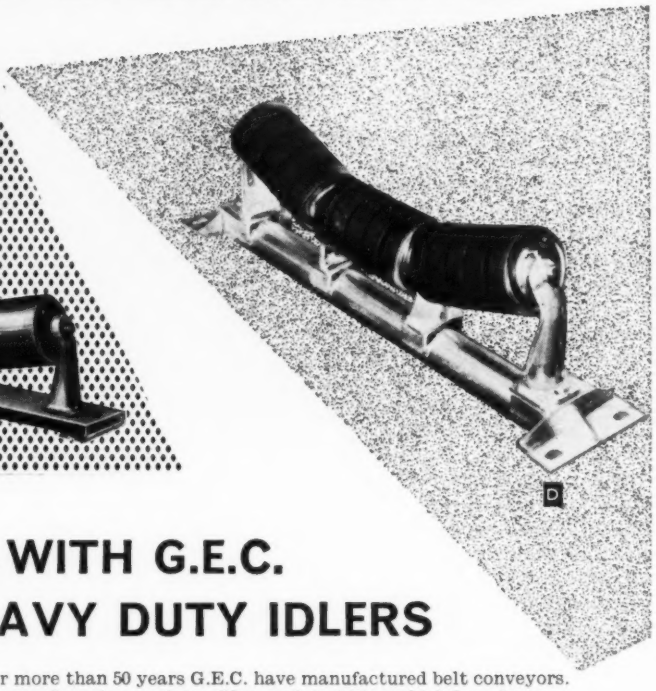
DUSTON, NORTHAMPTON (HEAD OFFICE) AND DAVENTRY, NORTHANTS





KEEP MATERIALS MOVING





A *6" diameter steel tube troughing idler and return idler.* Rollers are made from mild steel tube 6" outside dia. and 6 s.w.g. Strong, rolled section steel idler base board. All troughing idlers, except on reversible belts, provided with tilting blocks to aid training of the belt.

B *7½" diameter rubber disc return idler.* Tube carrying discs is solid drawn 6 s.w.g. and 3" outside dia. Discs are strengthened internally with mild steel ring and remain rigidly in position. Idler especially suitable where damp and sticky materials are handled.

C *6" diameter steel tube feeder idler.* Rollers are mild steel tube 6 s.w.g., and 6" outside dia. Contour of idler is designed so that material can be drawn freely from hoppers, etc., in relatively large quantities at low speeds.

D *6" diameter rubber disc impact idler.* Tube supporting discs is solid drawn ¼" thickness and 2½" outside dia. Idler base board is constructed for exceptionally arduous duties. Discs are designed to support and protect belt at loading points, cushioning shock when impact is severe.

. . . WITH G.E.C. HEAVY DUTY IDLERS

For more than 50 years G.E.C. have manufactured belt conveyors. Shown above is a variety of heavy duty idlers which have been developed through extensive research and 50 years' experience in the materials handling field. Important features of these idlers are—

SEALS designed to retain lubrication and exclude dust and abrasive matter—thus reducing friction and wear.
BEARINGS of a taper roller type capable of a continuous load of 1,500 lb., giving a high factor of safety.
SINGLE SHOT LUBRICATION on either outside bracket. All bearings need only infrequent lubrication.
BRACKETS cast from malleable iron, designed for maximum strength and resistance to shock loads and breakage.

All components are proportioned to allow for corrosion, abrasion and normal abuse. Special equipment can be supplied to meet abnormal conditions.

For further information write for Publication No. 2829

G.E.C.

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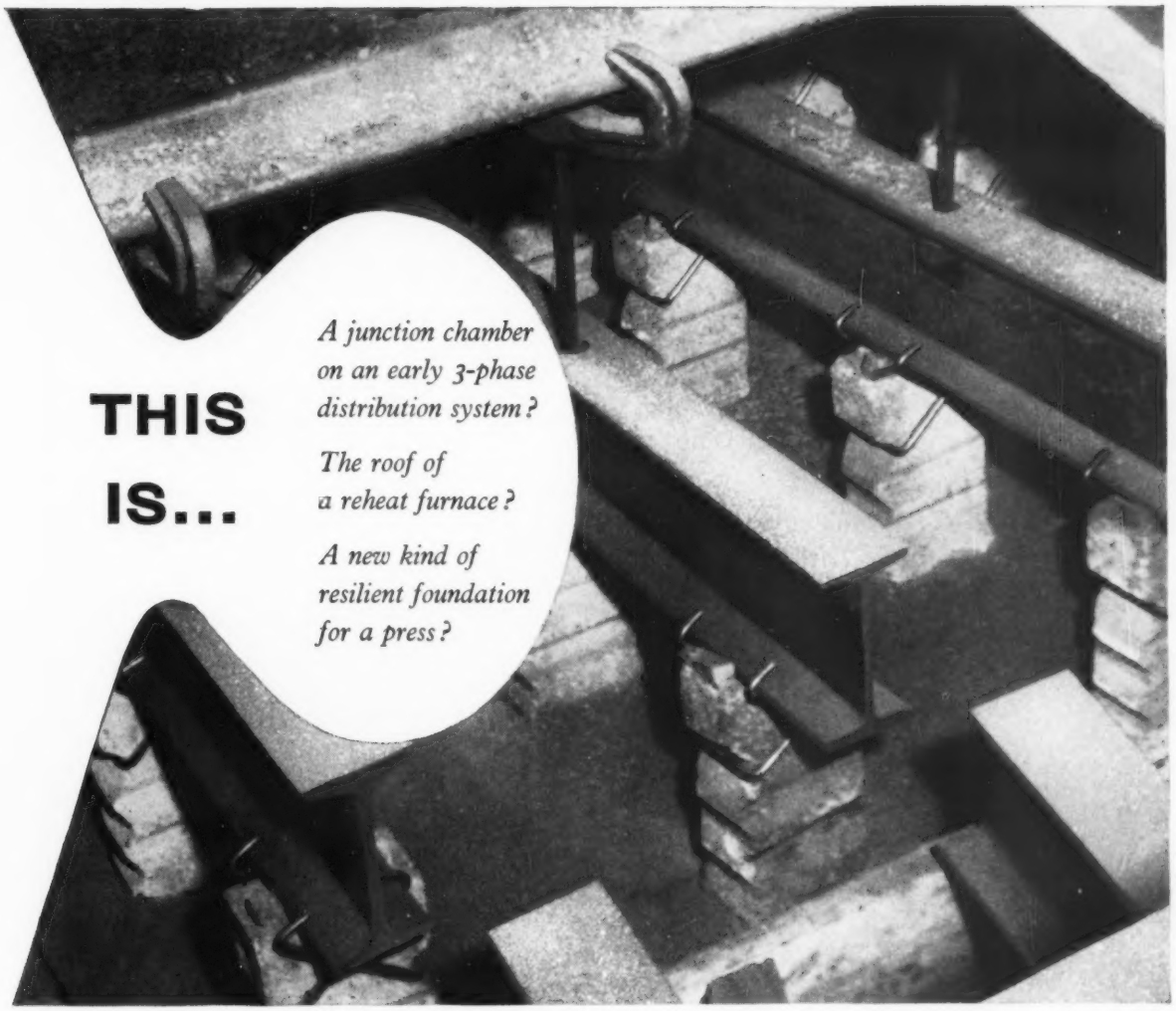
FRASER & CHALMERS ENGINEERING WORKS · ERITH · KENT
Designers and Manufacturers of Complete Materials Handling Plant.


THIS IS...
A junction chamber on an early 3-phase distribution system?
The roof of a reheat furnace?
A new kind of resilient foundation for a press?

Number two is the right answer. It is part of the monolithic roof of a reheat furnace at the Margam works of the Steel Company of Wales, immediately before the refractory was placed.

The special anchor bricks hanging from the R.S.J's are secured by heat-resisting clips and will provide support through the full thickness of the refractory lining.

This roof was cast in "TRI-MOR" High Temperature Castable. It is a good example of the type of furnace lining which "TRI-MOR" refractories and Morgan's whole-hearted belief in monolithic structures have made possible.

Some of the most useful of the "TRI-MOR" refractories are:—

TRI-MOR Standard Castable
TRI-MOR High Strength Castable
TRI-MOR High Temperature Castable
TRI-MOR High Temperature Mouldable
TRI-MOR Dense 'Guncrete'
TRI-MOR Insulating Castable
TRI-MOR Insulating 'Guncrete'

MORGAN Refractories Ltd

MORGAN REFRACTORIES LIMITED,
NESTON, WIRRAL, CHESHIRE. TELEPHONE: NESTON 1406

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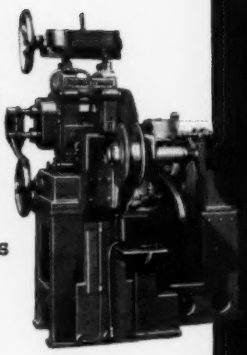
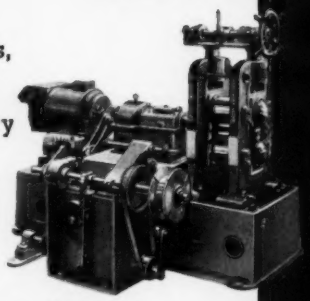
MAN AND MACHINES

Man perforce has become a designer of machines. Machines which will help him produce more goods more speedily. After he produces faster, he needs more accuracy, greater versatility. Especially if he happens to work in metals, in sheets or strips, wires or tubes.

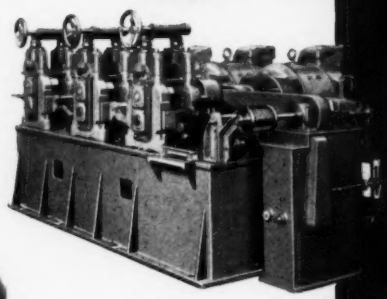
Robertson's equipment for reducing steel, aluminium and other metals into sheets, strips, wires and tubes, has developed consistently throughout the last half century in advance of man's requirements. Today, their high quality, precision machinery designed for sustained output is found throughout the world, wherever men are working metals.



A small Two-High Mill with rolls 4" diameter, 3" face.

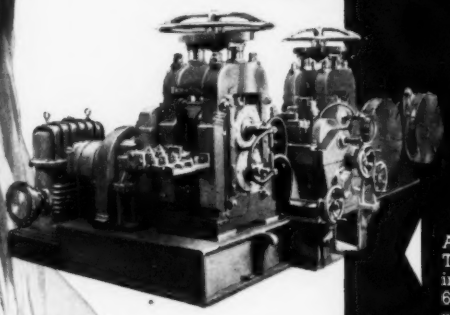


A Four-High Single Stand Mill with work rolls 2½" diameter, 4" face, and support rolls 6¼" diameter, 4" face.

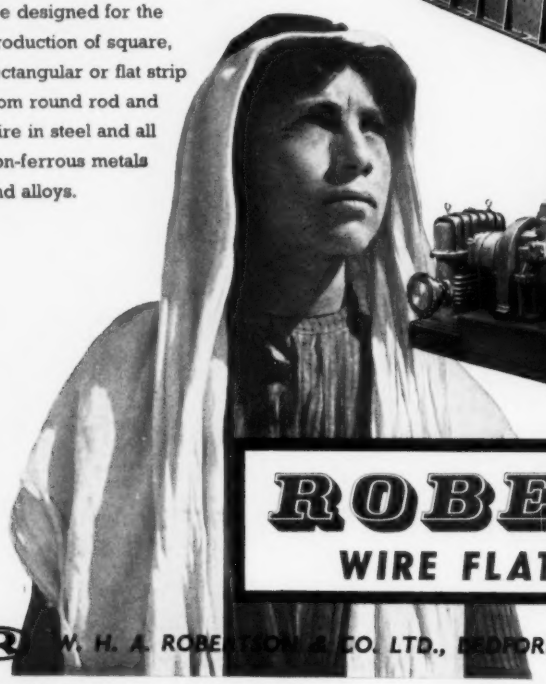


A Three-Stand Tandem, each stand individually motor driven and having rolls 6¼" diameter, 4" face.

The examples illustrated are from the range of Robertson Mills, which are designed for the production of square, rectangular or flat strip from round rod and wire in steel and all non-ferrous metals and alloys.



A No. 2 Tormanco Tandem, with roughing rolls 10" diameter, 6" face, and finishing rolls 8" diameter, 6" face



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WIRE FLATTENING MILLS

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LICENSEES FOR THE BUILDING OF SENDZIMIR COLD REDUCTION MILLS & PLANETARY HOT MILLS, HALLDEN GUILLOTINE AND ROTARY FLYING SHEAR MACHINES, AND TORRINGTON METAL WORKING MACHINERY. W T.303

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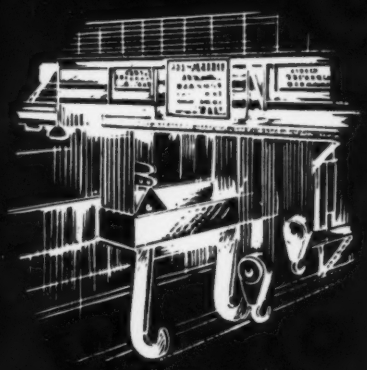
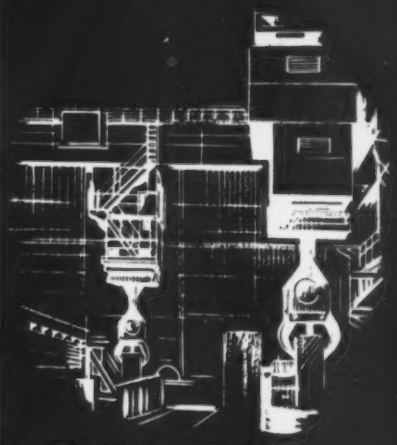
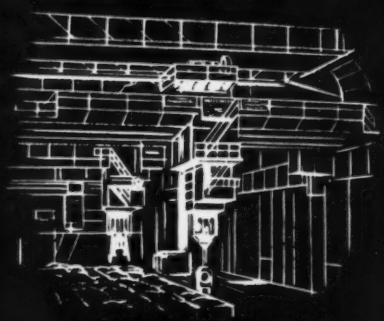


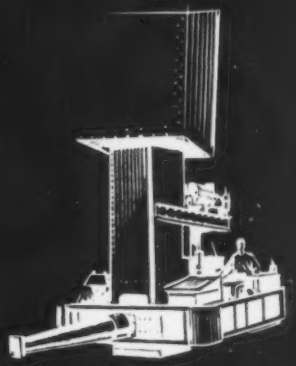
Ladle Cranes



Soaking Pit Cranes



Stripping Cranes



Charging Machines

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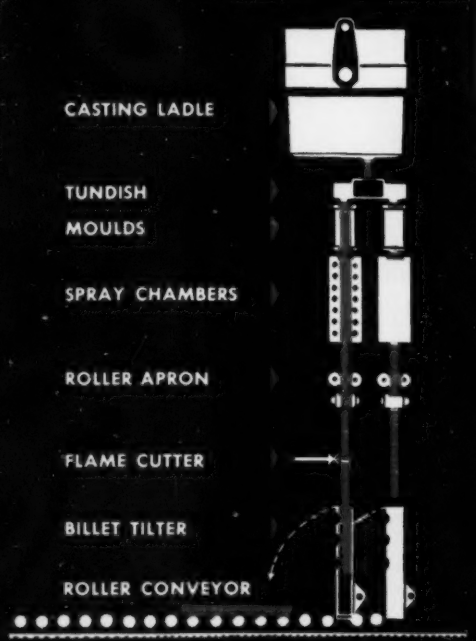

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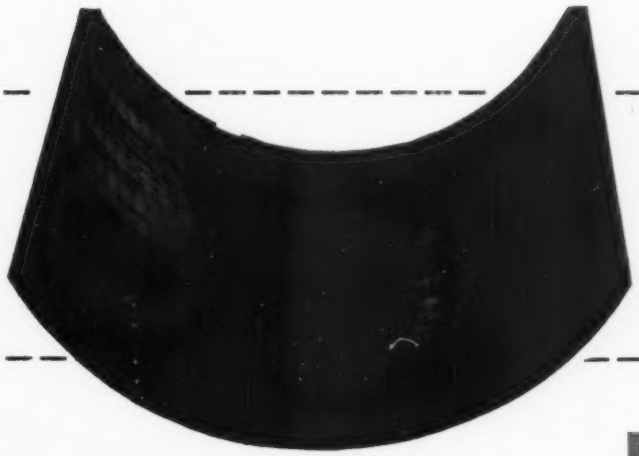

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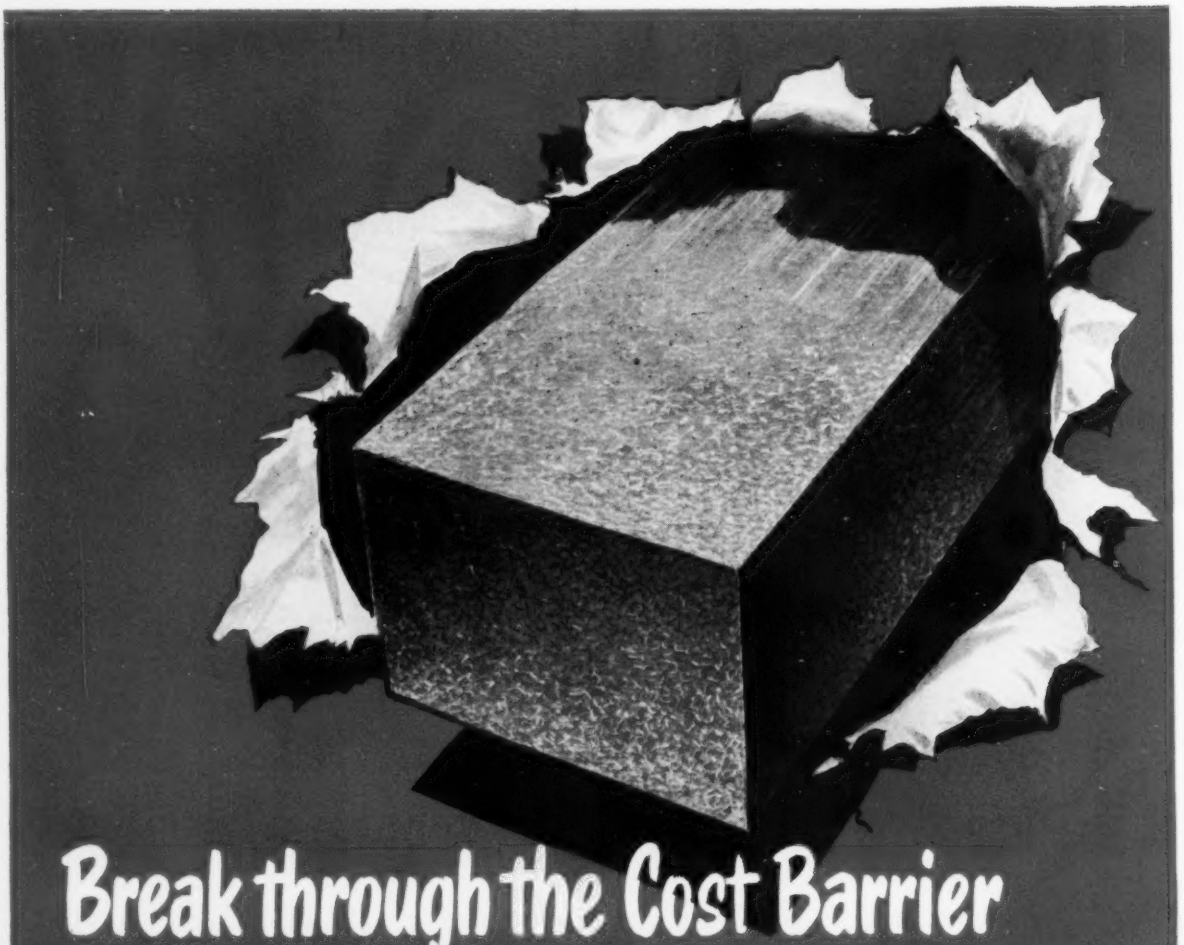

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Journal of The Iron and Steel Institute

AUGUST, 1959
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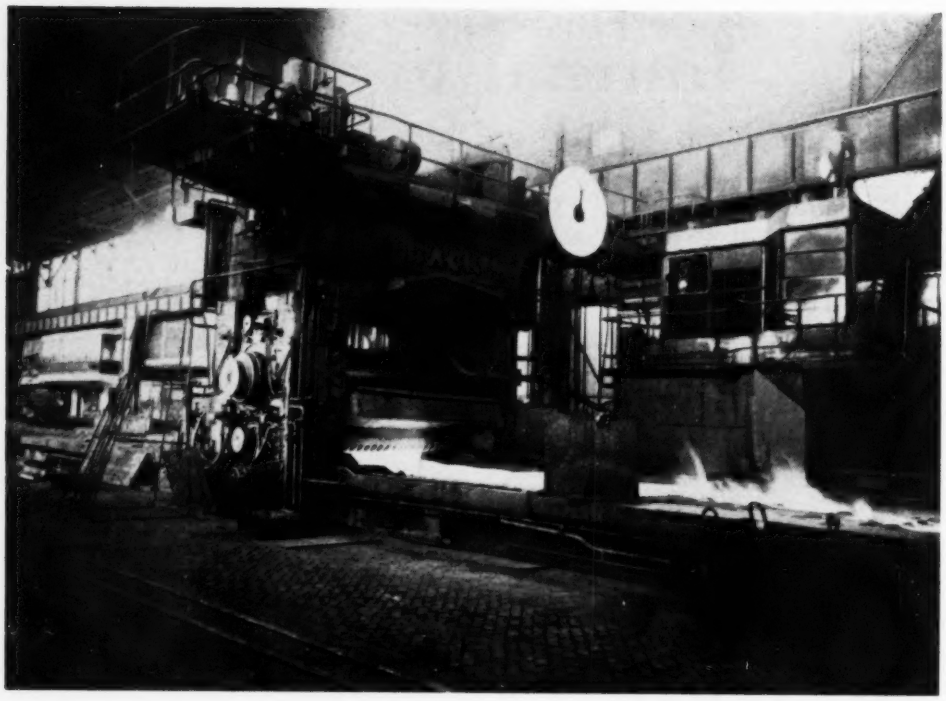

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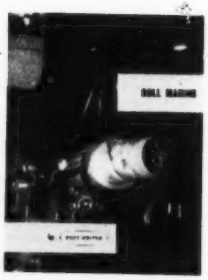

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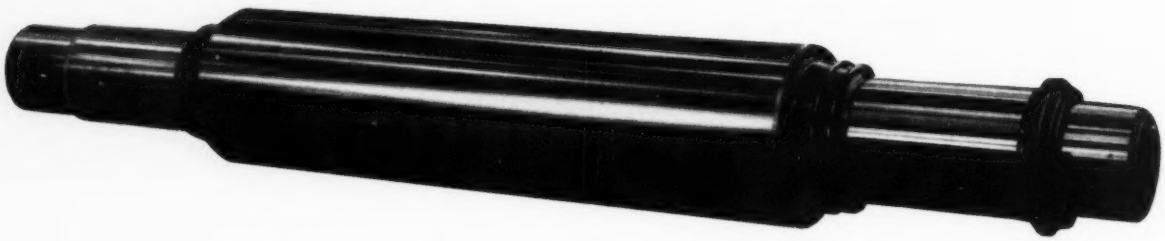

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W. F. GILBERTSON

William Frederick Gilbertson was educated at Lancing College, Sussex, and studied metallurgy at the University College of Wales, Swansea. In 1929 he entered the family tinplate firm at Pontardawe, where he remained for 13 years.

In 1942 he went to Baldwins Ltd, Swansea, as assistant general manager of the Kings Dock, Fairwood, and Elba tinplate works. On the merger of Baldwins Ltd with Richard Thomas and Company in 1945, Mr. Gilbertson moved to the West Wales Section office at Gorseinon to take charge as assistant general manager (tinplate) of the R.T.B. tinplate works. In 1949 he went to Redbourn works (Scunthorpe) and subsequently paid a four months' visit to the USA to study the iron and steel industry.

He was appointed assistant general superintendent at Ebbw Vale works in 1951, became assistant general manager a year later and deputy general manager in 1954. In 1957 he moved to the head office of the company in London as executive assistant to the managing director, and in 1958 he became, in addition, production controller.

Mr Gilbertson was elected vice-president of no. 4 division of the Iron and Steel Trades Employers' Association (Wales and Monmouthshire) in 1956, and in 1957 president of the Iron and Steel Trades Employers' Association. He is also deputy chairman of the Management Committee of the Tinplate Conference (1925) and in June 1959 he was elected to the board of Richard Thomas and Baldwins Ltd.

Mr Gilbertson has been a member of The Iron and Steel Institute since 1934, and in 1958 he was elected to the Council of the Institute.

W. F. Gilbertson
Member of Council

PROCEEDINGS OF

THE INSTITUTE

Reports of Council · Meetings · Notices

Report of the Annual General Meeting, 1959

The NINETIETH ANNUAL GENERAL MEETING of THE IRON AND STEEL INSTITUTE was held on Wednesday and Thursday, 6th and 7th May, 1959, at the Hoare Memorial Hall, Church House, Westminster, London, SW1. Mr C. R. WHEELER, C.B.E., the Retiring President, was in the Chair at the beginning of the Meeting, his place being taken later by Mr W. BARR, O.B.E., the new President.

Sessions were held at 10 A.M. and 2.30 P.M. on the Wednesday and at 10 A.M. on the Thursday. On the Wednesday evening a Dinner for Members was held at Grosvenor House, Park Lane, London, W1.

MINUTES OF PREVIOUS MEETING

THE MINUTES of the previous Meeting were taken as read, and were confirmed and signed.

WELCOME TO MEMBERS AND VISITORS

The **President** offered a cordial welcome to the large number of Members and visitors attending the Meeting. He welcomed in particular the visitors from overseas. It was a very good sign of the health of the Institute to see such a large number of Members and visitors present.

OBITUARY

The **President** referred to the loss which the Institute had suffered during the year through the death of Members. He referred in particular to the death of a great friend of the Institute, Sir William Larke. Sir William had been an Honorary Vice-President and a Bessemer Medallist, and at all times throughout his long life, and particularly between the wars, he had been a staunch friend of the Institute.

The Meeting, at the invitation of the President, stood in silence for a few moments as a token of respect.

PRESENTATION OF REPORT OF COUNCIL AND ACCOUNTS

The **President** then presented the Report of Council for 1958 and the Statement of Accounts, (*JISI*, 1959, vol. 192, May, pp. 8–12), and moved their adoption.

The Report and Accounts, he said, had been printed in the *Journal*, so he did not propose to go into it in detail.

The membership of the Institute, at 5 098, was 100 more than at the end of the previous year, and was the highest membership that the Institute had ever had. This was a matter for satisfaction.

During the year there had been a number of meetings, all of which had been well attended. He wished particularly to mention the rather noteworthy meetings which had been held in Belgium and Luxembourg last summer, and to thank the various representatives who were present at the Meeting today from Belgium and Luxembourg for the admirable arrangements which they had made and for the great pleasure which they had given to Members of the Institute.

There had also been Dr Allen's Hatfield Memorial Lecture on "The mechanical properties of the ferrite crystal," for which there had been a well attended and successful meeting; and meetings of the Engineers Group and Powder Metallurgy Joint Group, which had been well supported and extremely successful. The Engineers Group had held a meeting in Holland, and he wished to thank his predecessor in the Presidency, Mr Ingen-Housz, for the excellent hospitality which he had arranged for Members in Holland.

The *Journal* during the year seemed to him to have been exceptionally interesting. It had contained a wide variety of papers and, while dealing with specialized subjects, had contained a great deal for those who were generally interested in the industry. There was at the present time some 'jam' in regard to the publication of papers. The publication of papers now took longer than usual. The Council had yesterday agreed to measures to improve and accelerate the publication of papers. The *Journal* was to be enlarged, and it was hoped that by this means it would be possible to catch up with the flow of papers.

The Translation Service of the Institute had been successful. Nearly 400 translations had been issued during the year. This was, he felt, a very vigorous and useful service.

The Joint Library and Information Department required additional space for staff and storage of books. It had therefore been decided to build out at the back of the offices at 4 Grosvenor Gardens. In the meantime, they must sympathize with the staff in the difficulties to which it had been put and would be put until the alterations had been made. The staff deserved congratulations on coping with the service under very difficult circumstances.

Co-operation with other Societies had been maintained satisfactorily during the year. He need mention only a few; the British Iron and Steel Federation, the British Iron and Steel Research Association, the Affiliated Local Societies, The Institution of Metallurgists, and The Institute of Metals. He took the opportunity of congratulating The Institute of Metals on their Golden Jubilee this year. He had had the pleasure of attending the celebrations of that Golden Jubilee.

He felt that Members would agree that the Institute had had a vigorous and successful year.

Mr **W. Barr**, O.B.E., in seconding the motion for the adoption of the Reports and Accounts, said that this was the last occasion on which he would be giving an account of his stewardship; and he was sorry that the position was not a little bit rosier than it was in fact.

The accounts for 1958 were presented in the usual form. Increased expenditure was mainly due to increased activities and was reflected also in an increase in income. Nevertheless, the margin was, in his opinion, too small, since in order to maintain Reserves the Accumulated Fund had been reduced by £1 450; it now stood at £33 815. On the other hand, Reserves were now £3 750 and the House Fund amounted to £57 450; over £9 000 had been written off the lease of 4 Grosvenor Gardens during the last four years, so that the value of the lease had been written down to £35 500, which he though was very satisfactory.

It was appropriate to say a word about the future. It was anticipated that during the current year maintenance of Institute activities at full level would result in an excess of expenditure over income of the order of £4 000. On the other hand, in 1960, if the Council's recommendations were adopted, it was estimated that there would be a reasonable surplus after making full provision for all foreseeable expenditure, including the increased size of publications which was likely to be required.

In presenting the accounts of the Trust Funds, he need only refer to the Andrew Carnegie Research Fund. Some years ago there had been difficulty in using the money to advantage. During the last year or two, substantially more applications for grants had been received than could be met out of current income. They had, therefore, had to be considerably more selective in making awards.

He would be pleased to answer any questions that might be asked with regard to the Accounts.

The motion for the adoption of the Report of Council and the Statement of Accounts was carried unanimously.

MEMBERS' SUBSCRIPTION RATES

The **President** said that, as Members were aware, a circular had been issued to inform them that the Council had reluctantly come to the conclusion that an increase in subscription rates could not longer be postponed. The present rates had been in force since January, 1949.

Before proposing the required Resolutions, he wished to tell Members that it was the intention of the Council also to amend the regulations for Associate Membership. The effect of the proposed changes was that from 1st January, 1960, those whose age did not exceed 25 would continue to pay the same subscription as at present; those who were between the age of 25 and 28 would pay one guinea a year less than now; full Members who were over the age of 28 would be asked to pay one guinea a year more than at present.

The President then formally moved:

(1) That as from the 1st January, 1960, the annual subscription payable by ordinary members be increased to £5 5*s* and the annual subscription payable by associate members whose age exceeds 25 years be £3 3*s*.

(2) That as from 1st January, 1960, any ordinary member whose subscription is not in arrear may at any time before the 30th June in any year, or if elected after that date in the year, within one month of his election, compound for his subscription for subsequent years on payment according to the following scale:

Up to 29 years of age	£88
From 30 to 39 years of age	£79
From 40 to 49 years of age	£70
From 50 to 59 years of age	£61
From 60 years of age and upwards	£52

Provided that the passing of this Resolution shall not prejudice the validity of any composition duly effected according to the scale set out in Bye-Law 21 of the Institute's Bye-Laws.

Mr **W. Barr**, O.B.E., in seconding the resolution, said that Members would appreciate that there were very few Institutions which had not increased their subscriptions over the last ten years. He felt that this increase was timed very well, because it coincided with the new regulations under the Finance Act which enabled Members to get an income tax rebate on the subscriptions.

He wished to emphasize that in making these alterations they were catering for the younger Members. The fact that the younger Members were not being called upon to pay more was a point of advantage.

The resolution was carried unanimously.

PRESENTATIONS

Diploma of Honorary Membership

The **President** said that it was now his very pleasant duty to present a Diploma of Honorary Membership to their very good friend, Mr Ingen-Housz, who had been his predecessor in the Presidency and who was well known to the Members of the Institute and indeed to all those concerned with the iron and steel industry throughout the world. Mr Ingen-Housz had been, until recently, President of the Royal Netherlands Blast Furnaces and Steelworks Ltd. He had been a Member of the Institute since 1931. No man had given more enthusiastic support to the Institute than Mr Ingen-Housz, and it was with very great pleasure that he handed to him the Diploma of Honorary Membership.

Mr **A. H. Ingen-Housz** said that he had always served the Institute with the greatest pleasure, but he did not understand that he had served it so well that he deserved this very great honour.

Bessemer Gold Medal for 1959

The **President** announced that the Bessemer Gold Medal for 1959 had been awarded to Professor Count Bo Kalling of Sweden in recognition of his distinguished contributions to the physical chemistry of steelmaking and to the development of new processes in the making and treating of steel.

Professor Kalling was well known to all the Members of the Institute, if not personally, by repute, and those who had had the pleasure of knowing him personally took particular pleasure in this award. Professor Kalling's record was one of great distinction. In 1916 he had gone to the Domnarvet iron and steelworks, where he had been concerned with the electric reduction furnaces; in 1918 he had gone to Ferrolegeringar (Trollhättan) for research on ferro-alloys; in 1924 he had gone to Avesta Jernverks, where he had studied stainless steel; in 1931, he had been appointed Professor of Ferrous Metallurgy at the Royal Institute of Technology, Stockholm; in 1940, he had become Director of Research, Bolidens Gruv; and in 1945 he had been made Director of Research at Domnarvet for Stora Kopparbergs Bergslags, where he had stayed until 31st December, 1958.

Professor Kalling had made distinguished contributions to the physical chemistry of steelmaking, desulphurization, and the Kaldo process for refining phosphorus pig iron in a rotating furnace with the aid of oxygen. In 1957 he had received the Jernkontoret Gold Medal, with chain.

Professor Kalling was indeed a very distinguished and worthy recipient of the Bessemer Gold Medal.

Professor Count Bo Kalling said that it was a great honour and a source of immense gratification to be awarded this high distinction by the Institute.

He was happy and proud to be included among those metallurgists of outstanding merit both inside and outside the UK who had been awarded the Bessemer Gold Medal in the past.

Professor Kalling continued: "There is one point I should like particularly to stress. It is not for me to judge how far I have merited this great distinction, but I would strongly emphasize that hardly any results of importance are achieved from research nowadays that are not the outcome of close co-operation between a number of scientists working together. If my activities have led to any results of value, this is primarily because I have always had the advantage of receiving a powerful stimulus and encouragement from my skilled collaborators."

Sir Robert Hadfield Medal for 1959

The President announced that the Sir Robert Hadfield Medal for 1959 had been awarded to Mr A. Jackson, of the United Steel Companies Ltd, for his notable contributions to metallurgy and metallurgical education and, in particular, the advancement of the manufacture of steel from phosphoric pig iron by the open-hearth tilting furnace, culminating in the development of the Ajax oxygen process. Mr Jackson had recently been appointed adviser on steel to the United Steel Companies Ltd.

A great many Members of the Institute knew Mr Jackson very well. He asked Mr Jackson to receive the Medal, knowing that he had a host of friends within the Institute and that they wished to acknowledge the sound work that he had done.

WILLIAMS PRIZE

The President presented a Williams Prize for 1958 of £100 to Mr H. C. Child, of Jessop-Saville Ltd, for a paper on "Vacuum-melting of steels," which had been published in the *Journal* for December, 1958. The co-author, Mr G. T. Harris, was not eligible.

ANNOUNCEMENTS

The Secretary made the following announcements:

Andrew Carnegie Silver Medal for 1958

An Andrew Carnegie Silver Medal for 1958 had been awarded to Dr P. Vasudevan (formerly at the University of Manchester), who could not be present, for his report on "The kinetics of bainite formation in a plain carbon steel." (The co-authors Mr L. W. Graham and Mr H. J. Axon were not eligible.) The report had been published in the *Journal* in December, 1958.

Ablett Prize

No award had been made for 1958 of the Ablett Prize for a paper on engineering in iron and steelworks by an author under the age of 35.

Retiring Members of Council

In accordance with Bye-Law 10, the names of the following Vice-Presidents and Members of Council had been announced at the Autumn Meeting, 1958, as being due to retire at the present Annual Meeting:

Vice-Presidents: Mr E. T. Judge, Mr F. H. Saniter, and Dr C. Sykes, F.R.S.

Members of Council: Mr C. H. T. Williams, Professor A. G. Quarrell, Mr B. Chetwynd Talbot, Mr E. W. Colbeck, and Lt.-Com. G. W. Wells.

No other members having been nominated up to one month previous to the present meeting, the retiring Vice-Presidents and Members of Council were held to be re-elected.

Changes on the Council

Since the last General Meeting on 10th December, 1958, with effect from 6th May, 1959:

Honorary Treasurer:	Sir Julian Pode
Vice-Presidents:	Mr B. Chetwynd Talbot
	Mr C. H. T. Williams
Members of Council:	Major W. R. Brown, D.S.O.
	Mr T. R. Craig, O.B.E., T.D.
Additional Honorary Members of Council:	Dr N. P. Allen, F.R.S.
	Mr A. Jackson.

Since the last General Meeting on 10th December, 1958:

Honorary Members of Council:

Mr H. E. Gresham, President of the East Midlands Metallurgical Society, in place of Mr N. S. Angus.

Mr B. Ryan, President of Liverpool Metallurgical Society, in place of Dr C. Bodsworth.

Professor A. G. Quarrell (a Member of Council), President of the Sheffield Metallurgical Association, in place of Mr R. C. Baker.

Mr H. Bull, President of the Sheffield Society of Engineers and Metallurgists, in place of Mr E. D. Bool (deceased).

Professor R. Higgins, President of the Swansea and District Metallurgical Society, in place of Mr J. S. Caswell.

Dates of Meetings

Autumn General Meeting, 1959, in London, in conjunction with the Engineers Group, from 2nd to 4th December, to include discussions on forging and extrusion.

Twelfth Hatfield Memorial Lecture in Sheffield on 24th November, 1959. An invitation has been issued by the University of Sheffield, on the recommendation of the Lecture Committee, to Academician G. Kurdjumov, Director, Institute of Metallography and Metal Physics, Central Institute for Iron and Steel Research, Moscow.

Academician Kurdjumov has also been invited to lecture in London during the Institute's Autumn General Meeting, 1959.

Mr G. Adaryukov, Ironmaking Division of Gosplan and a member of the Editorial Board of Metallurg, USSR, has been invited in conjunction with BISRA, to give one or more lectures on blast-furnace practice during October, 1959.

Annual General Meeting, 1960, in London, from 4th to 6th May, to include a symposium on wide strip.

Joint Meeting with the Society for Analytical Chemistry and The Institute of Metals on the determination of gases in metals in May, 1960, possibly at the time of the Annual General Meeting.

Additional meetings of the Engineers Group include Junior Engineers in Glasgow on 28th and 29th October, 1959, by invitation of Colvilles Ltd and meeting in London in March, 1960, to discuss plate mills.

REPORT ON FUTURE ACTIVITIES

The President made the following announcements:

Special Meetings

There was to be a meeting on the North-East coast from 6th to 10th July. A circular had been issued, and an excellent programme had been arranged. They were very grateful to their friends on the North-East coast for making the preparations, and he was sure this would be an extremely pleasant and interesting visit.

From 29th May, 1960, by invitation of Associazioné Italiana di Metallurgia, there was to be a meeting in Italy. The invitation to hold a meeting in Italy had been accepted, it would begin in Milan on 29th May,

and there would also be visits to works near Turin, Genoa, Rome, and Naples. The programme was being devised so that Members might take a week or more according to their inclinations, and the programme would be expandable both geographically and by time.

In the autumn of 1961, by invitation of the Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers, it was planned to hold a meeting in the USA. The Council had given a good deal of thought to the possibility of holding a meeting in the USA. They were all agreed that it was most desirable from many points of view that they should have a meeting there. Accordingly it was with very real pleasure that he could tell Members that an invitation had been received from the Metallurgical Society of AIME. That invitation had been accepted.

They had close and friendly relations also with other societies in the USA and they hoped that some of those would be associated with the meeting. The Secretary of the American Iron and Steel Institute had said that that important organization would be very happy to assist in making the arrangements for plant and laboratory visits. He had offered also to participate in the planning of the technical programme. This was most satisfactory.

It was hoped that before long arrangements would also have been made for participation by the American Society for Metals.

This was only a preliminary intimation, but the Council had thought that it would be helpful to Members to know well in advance that arrangements were in hand for a meeting of the Institute in the USA in the autumn of 1961.

He was sure that all Members would wish to join with him in expressing appreciation to their American friends for their initiative in issuing their cordial invitation.

He added that Mr Barr and himself had accepted invitations to attend the meeting of the American Iron and Steel Institute which would be held in New York later this month.

Translation of Russian periodical ' Stal '

At the suggestion of DSIR the Institute had arranged to translate the Russian periodical ' Stal '. This was one of the leading journals in the world on iron and steelmaking. The translation of the issue for January, 1959, would be available in a few weeks and would be followed shortly by issues for the following months.

Staff appointments

The Council had, on the recommendation of the Secretary, approved the following changes in staff appointments:

Mr A. J. POST, at present Assistant Secretary, to be Deputy Secretary.

Mr H. F. CLEERE, at present Production Editor, to be Assistant Secretary and Executive Editor. In the latter capacity he would remain responsible to Mr E. R. MORT, Managing Editor.

Mr R. J. D. HARVEY, at present an Assistant Editor, to be Production Editor.

Proposed Memorial to Sidney Gilchrist Thomas

The Newport and District Metallurgical Society had made proposals to the Institute for a Memorial to Sidney Gilchrist Thomas, which the Council had considered and were proposing to support. Discussions would take place with the Newport and District Metallurgical Society as to details.

ELECTION OF MEMBERS

The scrutineers reported that the following candidates for membership had been duly elected:

Members

Akerlow, Earl V., (Oakland, Calif., USA); **Andrejević,** George, (Belgrade, Yugoslavia); **Arregui Sabarte,** Pedro, (Bilbao, Spain); **Ballantyne,** William Brian, L.I.M., (Germiston—TVL, South Africa); **Barr,** William, B.SC., (London); **Beaujean,** Robert, N.J.D. (Seraing, Belgium); **Beek,** Eugen Hans, (Düsseldorf, Germany); **Bentley,** James Tomlinson, M.SC. (TECH.), (Stafford); **Bernard,** John Wilson, (Saltburn); **Bersoux,** Henri C. E., (Grivegnée, Belgium); **Bianchini,** Osvaldo, DR. ING., (Genoa, Italy); **Bigas,** (Moselle, France); **Billings,** George Henry, (Stoke-on-Trent); **Boos,** Herbert, DIPL.ING., (Gelsenkirchen, Germany); **Boucheny,** Louis, (Paris, France); **Boutigny,** Robert, (Paris, France); **Brooks,** G. W., B.S.(MET.ENG.), (Houston, Texas, USA); **Browne,** Michael Woodthorpe, B.A., (London); **Bunyan,** John, M.B.E., B.SC.(ENG.), DIPL.R.T.C., (Brighton); **Capus,** J. M., B.SC.(HONS.), PH.D., (Birmingham); **Carion,** Robert Fernand Léon Auguste Willy, (Grivegnée, Belgium); **Carney,** Dennis J., B.S., (Duquesne, Pa., USA); **Chakrabarty,** Surja Kumar Sarma, (Bilaton); **Clark,** Sydney Townsend, O.B.E., (Luton); **Cochrane,** Derek Leslie, (London); **Colmant,** Gaston, (Paris, France); **Cooke,** Royce Douglas, (Sheffield); **Davies,** Joan, (Newport); **Decamps,** Georges, (Moselle, France); **Douglass,** Gordon, B.SC., (Middlesbrough); **Duncan,** Basil James, (Huddersfield); **Eade,** Graham Arthur, B.A., (Holmfirth); **Eastwood,** Donald Rex, (Sheffield); **Efjestad,** Dagfin, (Oslo, Norway); **Enegren,** Uno, (London); **Epure,** Stefan, M.S.(MECH. ENG.), PH.D., (Los Angeles, Calif., USA); **Estruch-Subirana,** Benjamin Joseph, PH.D. (Billingham); **Evans,** Edrydd, (London); **Firket,** A., (Seraing, Belgium); **Flavell,** H., (Loughborough); **Frazier,** Robert H., B.S.(MET.ENG.)., (Youngstown, Ohio, USA); **Freeze,** Jonathan Douglas, (London); **Frehser,** Josef, DR.ING., (Kapfenberg, Austria); **Fromson,** Howard Arthur, (New York, NY, USA); **Geijer,** Ernst, (Riverton, NJ, USA); **Gendle,** James William, A.M.I.MECH.E., (Middlesbrough); **Ghose,** Aurobindu, (Glamorgan); **Gribble,** Donald Cecil Edward, B.SC., (London); **Grinter,** Brian Joseph, L.I.M., (Swansea); **Guhathakurta,** Manoj Kumar, (Calcutta, India); **Gyöngyösi,** Nandor, (Budapest, Hungary); **Halmos,** George T., (Toronto, Ontario, Canada); **Hammarlund,** Per-Erik, M.SC., (Västerås, Sweden); **Hamou,** Albert, (Rabat Chellah, Morocco); **Harr,** Albrecht, DR.ING., (Dortmund, Germany); **Harris,** Max, B.SC.(ENG.), A.I.M., (Johannesburg, South Africa); **Hauttmann,** Hubert, DR.ING., (Linz-Donau, Austria); **Heck,** Klaus J. E., DIPL.ING., (Zurich, Switzerland); **Heynert,** Gerhard, DR.ING., (Duisburg-Ruhrort, Germany); **Hicks,** Lawrence Whitby, A.MET., A.I.M., (Sheffield); **Hocking,** James Samuel, (Dagenham); **Holland,** Cecil Bernard, B.SC.(ENG.)HONS., (Wednesfield, nr. Wolverhampton); **Homes,** Georges A., DR.-ÈS-SC., (Charleroi, Belgium); **Hudson,** William John, B.SC.(ENG.MET), A.R.S.M., (Hamilton, Ontario, Canada); **Hughes,** John Grosvenor, A.M.I.E.E., (Stafford); **Jaffe,** John Walter Paul, M.A., A.M.INST.C.E., A.M.I.STRUCT.E., (London); **Jones,** Eric Gwynne, (Llanelly); **Jones,** T., (Grangetown, Middlesbrough); **Kalling,** Ulf, (Stockholm, Sweden); **Labrum,** E., C.G.I.A., A.I.M., (Dagenham); **Landuyt,** Miriam M. (Mrs), B.S., (Peoria, Ill., USA); **Lanzing,** Paul Willem Anton, (IJmuiden, Holland); **Leslie,** William Cairns, B.MET.E., M.SC., PH.D., (Monroeville, Pa., USA); **Lewis,** John T., (Bournemouth, Hampshire); **Love,** Leonard Fernley, A.M.I.E.E., (Stafford); **McKelvie,** James, B.SC., (Normanby, nr. Middlesbrough); **Malcolm,** Victor, M.A., (London); **Matthews,** Bernard Charles, A.I.M., (Walsall); **Meier,** Walter, DIPL.ING., (Zürich, Switzerland); **Meliere,** Carlo, (Genoa-Cornigliano, Italy); **Metcalfe,** Kenneth, B.SC., A.R.I.C., F.I.M., (Edgware); **Montero Aires,** Jose, (Trubia, Spain); **Mukerji,** Mohi Mohon, M.S., (Drug, India); **Newkirk,** John B., D.SC., M.S., B.MET.E., (New York, NY, USA); **Nicolelis,** Efthemius H., M.S., (Brooklyn, NY, USA); **Nisbet,** D. R. M., A.R.C.S.T., A.M.I.E.E., (Scunthorpe); **Noel,** Emmanuel A. L., (Seraing, Belgium); **Palazzi,** Aurelio, D.CHEM., (Genoa-Cornigliano, Italy); **Patrick,** Terence George, (Thorneliffe, nr. Sheffield); **Percival,** Frederick William, (Dowlais); **Plockinger,** Erwin, DR.MONT., (Essen, Germany); **Ponniah,** T. Frederick, B.E.(MECH.ENG.), (Port Kembla, NSW, Australia); **Pouncy,** John, B.SC., (Port Talbot, Glamorgan); **Redway,** Paul Warwick, T.D., J.P., (Saltburn-by-the-Sea); **Reside,** William, (Glasgow); **Richards,** Robin Owen, D.S.C., M.A., ASSOC.M.INST.GASE., M.INST.F. (London); **Ridout,** Philip James, A.I.M., (Birmingham); **Roze,** Horace Carlyle, (London); **Saunders,** Peter, (London); **Schees,** Pierre, (Tilleur, Belgium); **Shaw,** Bernard, (Basildon, Essex); **Sheriff,** Graham, L.I.M., (Nottingham); **Singh,** Balwant, B.A., (London); **Skelly,** George Arthur, (London); **Slack,** Linton Lewendon, (Pontypridd, Glamorgan); **Speich,** Gilbert Robert, D.SC., (Monroeville, Pa., USA); **Stanistreet,** Thomas James, (Stoke-on-Trent); **Starratt,** Fred Weston, Jr., M.A., B.S., (New York, NY, USA); **Stassin,** G., (Charleroi, Belgium); **Stewart-Jones,** Edward Thorold, (Redditch); **Stitt,** Robert Ellsworth, B.SC.(MET.), (Munhall, Pa., USA); **Stouff,** René Charles, (Paris, France); **Studley,** Thomas, B.SC., A.I.M., A.INST.P., (Newport); **Syme,** David McDonald, B.SC., A.R.C.S.T., (Motherwell); **Varley,** Philip Copley, M.B.E., T.D., M.A., F.I.M., (Gerrards Cross); **Vayda,** A., (Paris, France); **Vecchietti,** Luigi, (Varese, Italy); **Vickers,** John Bernard, M.MET., F.I.M., (Sheffield); **Vilain,** Arthur J. E., (Mons, Belgium); **Warebrook,** Reginald Harold, A.M.I.MECH.E., (Haulbowline, Co. Cork, Ireland); **White,** Carol Henry, M.MET., A.I.M., (Sheffield); **Willems,** Jacob, DIPL.CHEM., DR.RER.NAT., (Düsseldorf, Germany); **Williams,** Alan Edwin, A.M.I.E.E., (Bedford); **Willmer,** Stanley Wood, (Corby); **Wilson,** Tom Maddison, (Thornaby-on-Tees); **Wood,** Harleston R.,

(Conshohocken, Pa., USA); **Wrangham,** Cuthbert Edward, C.B.E., B.A., (Stockton-on-Tees); **Zehringer,** Pierre, (Paris, France).

Associate Members

Alexander, James David, (Sheffield); **Baker,** Robert, (Sheffield); **Bentley,** Kenneth Paul, (Cambridge); **Brown,** Alan John, B.A.(HONS.), (Corby); **Brown,** Henry Tarbitt, (Nottingham); **Daniell,** E. R., (Chesterfield); **Davies,** Byron, (Swansea); **Davies,** Dillwyn Morgan, B.SC., (Oxford); **Evans,** Christopher John Gillingham, (Cambridge); **Gittins,** Anthony, (Sheffield); **Green,** Ronald Pattinson, B.SC., (London); **Hogg,** James Sinclair, B.SC., (Glasgow); **Johnson,** John Henry, (Bilston); **Lyne,** Michael, (Swansea); **Makower,** Michael Stanley, (Cambridge); **Morris,** Alan William Henry, (Birmingham); **Nicholas,** Colin John, (Swansea); **Osborne,** Francis John, B.SC., (Cambridge); **Paramahamsa,** Nishtala Ramakrishna, B.SC., B.TECH. (HONS.)(MECH.ENG.), (Fort Vizianagram, S.E. Rly., India); **Parker,** Roger Hill, B.A., (Shotton); **Patrick,** Brian, (Nottingham); **Roberts,** Colin, (Swansea); **Romer,** Hugh Frank, (Cambridge); **Rothwell,** George Peter, (Cambridge); **Shillitoe,** David Rickman, (London); **Sidey, Paul,** (Swansea); **Spira,** Terence Anthony, (Hayes); **Tuffrey,** Anthony, (Nottingham); **Wright,** Michael George, (Swansea); **Wynblatt,** Paul, M.SC., (Rehovot, Israel); **Zia-ul Haq,** A. H. M., B.SC.(ENG.MET.), (Lahore, West Pakistan).

The President declared the applicants duly elected; they numbered 128 Members and 31 Associate Members and the total membership of the Institute was now 5 052.

INDUCTION OF NEW PRESIDENT

The retiring **President** (Mr C. R. WHEELER, C.B.E.), before inducting the President-Elect into the Chair, spoke of him as an outstanding metallurgist, well known throughout the UK and the world.

He added: "I say with sincerity that I know that, in electing Mr Barr as our President for this year, we have a man of distinction, sterling worth, and charming personality."

VOTE OF THANKS TO THE RETIRING PRESIDENT

The new **President** (Mr W. Barr, O.B.E.), said that his first Presidential duty was a very pleasant one; it was to thank the retiring President, Mr Wheeler, for all that he had done. He had come to the end of a highly successful Presidential year, and had been indefatigable in his attention to the day-to-day work of the Institute, and had travelled far and wide representing the Institute at various important technical functions both at home and abroad.

The vote of thanks was seconded by the **Hon. R. G. Lyttelton** (*Past-President*) and carried with acclamation.

PRESIDENTIAL ADDRESS

The **President** then delivered his address entitled "Progress in the development and application of steel plates" (*JISI*, 1959, vol. 192, June, pp. 109–117).

A vote of thanks to the President for his address, proposed by Mr **D. A. Oliver** (*Vice-President*) and seconded by Mr **E. T. Judge** (*Vice-President*) was carried with acclamation.

TECHNICAL SESSIONS

A number of papers were presented and discussed at technical sessions on both days of the meeting. Details were given in the News section of the April issue of the *Journal* (p. 390). Reports of the discussions will be published in the *Journal* later this year.

DINNER FOR MEMBERS

The annual Dinner for Members was held at Grosvenor House, Park Lane, London, W1, on the evening of Wednesday, 6th May, and was attended by about 1 500 Members and guests. The toast of "The Iron and Steel Institute" was proposed by the Rt Hon. J. S. MACLAY, P.C., C.M.G., Secretary of State for Scotland; the President replied. The health of the guests, proposed by SIR CHARLES GOODEVE, O.B.E., D.SC., F.R.S. (Vice-President and Director of the British Iron and Steel Research Association) was replied to by His Excellency Dr A. J. Clasen, DR. ING., D.SC., (Ambassador to the Grand Duchy of Luxembourg).

Stal in English

The publication of *Stal in English* has been delayed owing to the printing dispute. *Stal in English* (2) appeared at the beginning of September; the main contents are given below. Full details of subscriptions to *Stal in English* are available from the Secretary of the Iron and Steel Institute.

IRON AND STEELMAKING

Gas-flow distribution in the blast-furnace stack. *N. N. Babarykin.*
Operating experience with a covered ore-fines stockyard. *P. A. Gorbunov and A. S. Sazonov*
Hot-metal ladle operation in the Magnitogorsk blast-furnace shop. *G. I. Gerasimov et al.*
Converting a 400-ton open-hearth furnace to sulphurous crude-oil firing. *F. D. Voronov, E. I. Dikshtein, K. A. Zuts, and A. G. Trifonov.*
Heterogeneity in transformer-steel ingots. *N. F. Dubrov, et al.*

ELECTROMETALLURGY

Improved methods of melting transformer metal. *N. V. Keis, D. G. Zhukov, and A. M. Khizhnichenko.*
The mechanism of the influence of magnesium on steel. *I. A. Andreev and A. Ya Borisov.*

ROLLING AND TUBE PRODUCTION

Rationalized roll-pass design in continuous billet mills. *M. I. Boyarshinov and I. P. Shulaev.*
Increasing the life of cogging-mill rolls by work-hardening. *L. V. Andreyuk.*
Flame scarfing stainless and heat-resistant steels. *D. I. Gololobov and B. M. Litvinov.*

METAL SCIENCE AND HEAT TREATMENT

Generating a protective atmosphere for bright annealing. *F. I. Zenchenko and Ya. A. Peker.*
Tellurium in steel. *Ya. E. Gol'dshtein.*
Structural stability and properties of heat-resistant chromium steels. *A. V. Merlina and A. M. Borzdyka.*

ECONOMICS AND ORGANIZATION

Labour productivity at iron and steelworks. *B. B. Rozin and R. S. Geifman.*

METALLURGICAL HEATING TECHNIQUES

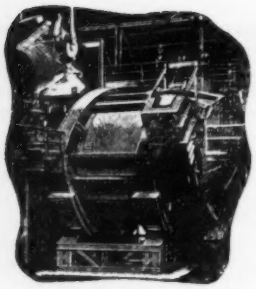
Experience in the use of recuperative soaking pits. *M. E. Boikov et al.*
Aerodynamics of bell-type furnaces for bright annealing. *V. I. Mitkalinny and N. G. Molchanov.*

FOUNDRY PRACTICE

Medium-sized ingot moulds in spheroidal graphite cast iron. *A. V. Bobrov and A. A. Sinitayn.*

POWER AND PLANT

Cleaning the top gas from ferromanganese production. *M. M. Zaitsev et al.*



PAPERS AND REPORTS ON RESEARCH AND PRACTICE

Metallurgy · Plant Operation · Research

Further experience with the Kaldo process

By Bo Kalling and Folke Johansson

AT THE MEETING in Liège last year that was jointly arranged by The Iron and Steel Institute and the Belgian CNRM, the authors had the opportunity to present the results by then attained with the Kaldo process. Since then, however, new experience has been gained which may well account for again giving a summary of the features of the process, its latest development and future prospects. Regarding the ordinary operations practice of the process, reference is made to previous publications[5, 6] from which Figs. 1 and 2 have been taken.

HEAT ECONOMY

Characteristic of the process and of great importance for its economy is the possibility of burning inside the furnace all or at least the main part of the carbon monoxide generated during the blow. To utilize the heat developed in this way for the process, the furnace has to be kept in rapid rotation so as to achieve good heat transmission to the bath and, at the same time, protect the furnace lining against overheating. At oxygen processes in stationary furnaces it has not proved possible to utilize the combustion heat of the carbon monoxide for the process itself to any considerable extent.

The percentage analysis of the pig iron normally used at Domnarvet is as follows:

C	Si	Mn	P
3·6	0·3	0·7	1·85

When cooling by ore, 140 kg/ton of pig iron are usually added at Domnarvet in addition to about 150 kg of lime, necessary for the slag formation. The ore contains 60%Fe which increases the output of steel by about 80 kg/ton of pig iron. These results

Paper presented to the Annual General Meeting of the Institute, May, 1959.

Professor Kalling was until this year Director of Research at Stora Kopparbergs Bergslags AB and Mr Johansson is a member of the research staff.

SYNOPSIS

The Kaldo process has been in use at Domnarvet since May, 1956, and some 200 000 tons of ingots have been produced, mainly from pig iron containing 1.85%P, although a low-phosphorus pig iron has also been satisfactorily treated. As most of the carbon monoxide produced during the blow is burnt to carbon dioxide, heat economy is considerable and much iron ore and scrap can be charged during the run. Iron losses in the slag are very low, and the iron yield is about 90%; desulphurization is very effective and nitrogen pickup is extremely small. Control of the process is very accurate and tapping temperature and chemical analysis can be kept within close limits, so that recarburization is not necessary. Refractory life has been improved from 50 heats per lining to 70, longer life is hoped for, but in spite of the high cost of refractories the cost of Kaldo steel has been brought below that of basic Bessemer. The steel is of good quality and several customers find it highly suitable for deep-drawing; at Domnarvet itself the material is largely used for ship-plates, for which its high impact-ductility, even after strain ageing, and its good weldability make it extremely suitable. 1727

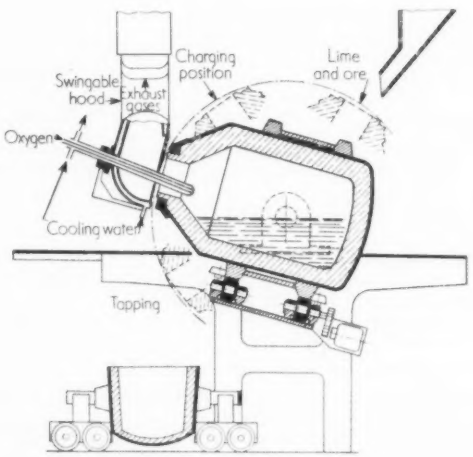


Fig. 1—30-ton Kaldo furnace

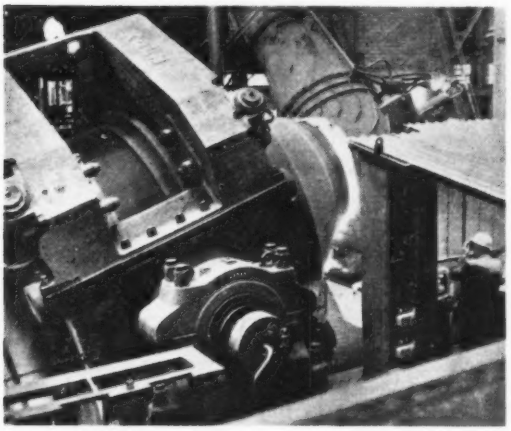


Fig. 2—Kaldo furnace in operation at Domnarvet

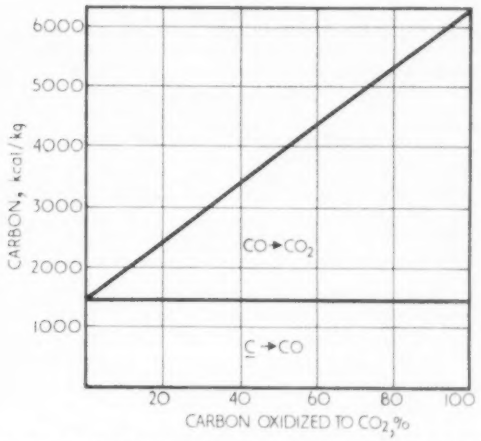


Fig. 3—Available heat from carbon by refining with oxygen

are obtained when burning on an average 75% of the carbon monoxide inside the furnace.

When refining low-phosphorus pig iron containing about 4·0%C and 1·0%Si, at least the same quantity of ore can be added. Using scrap instead of ore, it has been found possible to add 600 kg/ton of pig iron.

The quantities of cooling agents which can be added under different conditions during the course of the blow, will be examined later.

Table I gives the theoretically generated heat per kilogram of C, Si, Mn, P, and Fe. Deductions have been made for the quantity of heat that is bound in the reaction products and therefore cannot be utilized for the process. The composition and the assumed temperature of the reaction products have also been indicated in the table. The table shows that four times as much heat can be utilized, if the carbon is burnt to CO_2 instead of CO inside the furnace.

Figure 3 shows the available heat, if the carbon monoxide is only partially burnt. This diagram is instructive, as a comparatively small combustion of CO to CO_2 is also achieved in other pneumatic processes and, moreover, in the Kaldo process it is not always possible to burn the carbon monoxide completely. It is quite apparent from the diagram how much heat the carbon combustion gives, if the composition of the exhaust gases is known.

Table I also shows that it is possible to utilize 37%

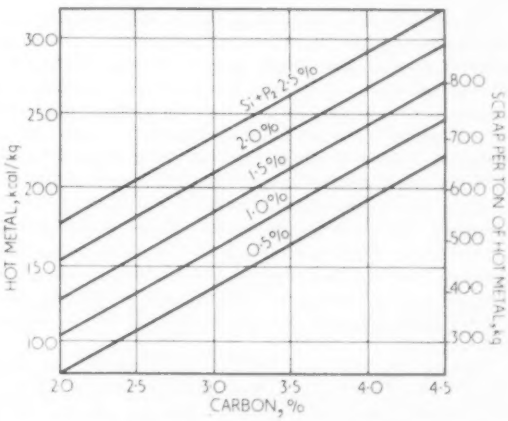
TABLE I

Available heat from hot-metal component

Element	Reactions	Gas temp., C	Slag temp., C	Available heat, kcal kg elem.
C	$\underline{C} \longrightarrow CO$	1 600	...	1 470
	$\underline{C} \longrightarrow CO_2$	1 800	...	6 240
Si	$\underline{Si} \longrightarrow 1{\cdot}5CaO.SiO_2$	...	1 500	5 235
	$\underline{Si} \longrightarrow 3CaO.SiO_2$	...	1 600	3 827
P	$\underline{P} \longrightarrow 3CaO.P_2O_5$	...	1 600	4 824
Mn	$Mn \longrightarrow MnO$	...	1 600	1 206
Fe	$Fe \longrightarrow FeO$	...	1 600	650

more heat from the silicon content of the pig iron if the silica formed is bound in a low-basic slag which is tapped off at 1500° C than if the silica remains in the furnace till the end of the blow when the slag has to be more basic to bind phosphorus and sulphur. It has proved advantageous also from other points of view to tap off the pre-slag after oxidation of the silicon, when high-silicon pig iron is used in the Kaldo process. Thus a better heat efficiency can also be expected from the silicon than in other processes where no pre-slag is tapped off.

In Fig. 4 a summary has been made of the heat available for scrap or ore melting when using pig iron of different compositions. In preparing this diagram, it has been assumed that 80% of the carbon has been burnt to CO_2 which, according to Table I, corresponds to about 5 300 kcal/kg C. For silicon, the estimate is 4 600 kcal/kg which is somewhat above the average of the two values, computed according to Table I, and consequently implies that the major part of the



Basis for calculations: 5 286 kcal/kg C, 4 600 kcal/kg (Si + P_2) hot-metal temp. 1 300° C; steel temp. 1 600° C; scrap equivalent 332 kcal/kg

Fig. 4—Available heat for melting scrap and ore in the Kaldo process

TABLE II

Example from desiliconizing in the Kaldo furnace

Charged: 25 700 kg of hot metal
500 ,, ,, lime
165 ,, ,, sinter (cold furnace)

Blowing: 440 nm³ O_2 for 9 min
Extra rotation without any O_2 for 1 min

Metal analysis, %	C	Si	Mn	P	S	°C
Before blowing	4·08	0·91	1·08	0·160	0·041	1 190
After ,,	3·76	0·00	0·71	0·194	0·023	1 480

Slag analysis, %	CaO	MgO	SiO_2	P_2O_5	MnO	FeO
After blowing	46·7	4·49	38·2	0·03	7·2	1·74

silica shall be removed in a pre-slag. Also for phosphorus, 4 600 kcal have been reckoned with, which is lower than the theoretical but more in accordance with practical experience, since a high-phosphorus slag always demands a certain excess of lime to effectively bind the phosphorus.

After summing up the available heat from the carbon, silicon, and phosphorus reactions, deductions have been made for the heat that is required to raise the temperature of the melt from 1 300° to 1 600° C. Furnace losses and some other losses, for instance those due to impurities in the lime, have not been taken into account. These are, however, partly covered by the surplus heat from the oxidizing of manganese and some iron. When applying the diagram to heats in the 26-ton Kaldo furnace at Domnarvet, where cooling is mainly done by scrap, it has proved that the heat available for scrap melting is about 20 kcal lower than the calculated heat, corresponding to about 60 kg scrap/ton of hot metal. However, in a 100-ton furnace designed for quick charging of scrap and other additions, the expected surplus heat can be obtained from the diagram.

Applied to pig iron with 4·0%C, 1·0%Si, and 0·22%P, which is refined to steel with 0·1%C and 0·02%P, a surplus heat of 223 kcal/kg of pig iron is obtained according to the diagram which is sufficient for melting 670 kg of scrap per ton of hot metal.

When cooling by ore, it is difficult to burn the carbon monoxide as completely as when cooling by scrap. When the ore is charged at the start or in a few large additions during the blow, the generation of carbon monoxide becomes so intense that it will be difficult to combust all the gas. A continuous feeding of ore, however, gives a more uniform reaction and improves the heat economy. As soon as the equipment for a more continuous ore charging now being designed, can be brought into use, we expect a better heat economy also at ore cooling. To enable the use of rich and at the same time cheap cooling ore, a practical charging device ought also to be adaptable for charging of ore concentrates. The large quantities of scrap that can be melted in the Kaldo process allow remelting of all scrap in integrated steel plants, and special steel furnaces for scrap melting are therefore not required.

The saving obtained by the additions of ore and scrap respectively, can be calculated approximately in the following way. If cooling is done by ore at a price of £6 per ton of iron content and the price of hot metal is supposed to be £20 per ton, a gain of about 28*s* per ton of pig iron is obtained at a reduction of 100 kg Fe from iron ore. On the other hand, if

TABLE III

Calculated metal yield for different operation conditions

Hot metal type		High phos.	Low phos.
Metal analysis		3·5%C 0·3%Si 1·8%P	4·0%C 1·0%Si 0·2%P
Scrap cooling			
Per 1 000 kg of steel:	Hot metal, kg	670	670
	Scrap ,,	412	398
	Total ,,	1 082	1 068
Metal yield		92·4	93·6
Ore cooling			
Per 1 000 kg of steel:	Hot metal kg	1 030	1 010
	Fe from ore ,,	88	85
	Total	1 118	1 095
Metal yield		89·4	91·3

remelting of scrap in another furnace would cost, for instance, £2 per ton, 27*s*/ton of pig iron can be saved by remelting 670 kg/ton of pig iron in the Kaldo furnace. The amount of cooling agents that can be added is, consequently, of great economic importance. Naturally, it depends on the raw material costs and the local conditions which cooling agent gives the greatest saving, and it must therefore be calculated in each case.

The combustion of carbon monoxide and the utilization of the combustion heat inside the furnace also involve another essential advantage with the Kaldo process. Determining for the size of the gas-cleaning equipment is the maximum heat content in the exhaust gases, i.e. the total amount of chemical and physical heat. At Domnarvet, this maximum heat during the blow amounts at present to about 180 000 kcal/min and can be reduced to a much lower value when cooling with scrap or continuous feeding of ore. This will give a much cheaper gas-cleaning system than is required when using a stationary furnace of the same capacity.

CONTROL OF THE PROCESS

The blow can be followed and adjusted by various simple observations and measures. The iron content and the viscosity of the slag can, among other things, be affected by changing the direction of the oxygen lance. If the oxygen jet is blown more vertically against the bath, the oxidation of iron is increased which makes the slag more fluid.

The speed of rotation also has a great influence on the run of the process. When using a high speed, the content of iron oxides becomes lower and the slag more viscous. A decrease in speed has the opposite effect. To reach a carbon content of less than 1% before the main part of the phosphorus is oxidized, it has been found necessary in the Domnarvet practice to maintain a high speed which delays the melting of the slag. Top speed is also desirable towards the end of the blow to keep down the iron content in the final slag.

A relatively dry slag is used during the main part of the blow; only before the slag is tapped the iron content is adjusted to get the slag sufficiently fluid.

Valuable information as to the condition of the slag inside the furnace is obtained by observing the load on the rotation motors during the blow, as a marked decrease on ampere meter occurs simultaneously with slag melting.

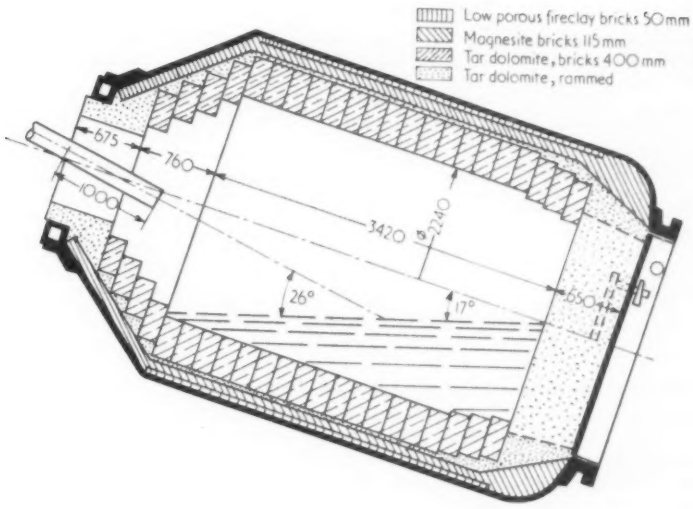


Fig. 5—Present lining of the Domnarvet Kaldo furnace

A good indication of how the reactions are proceeding is further obtained by measuring the temperature in the exhaust-gas pipe. As only carbon gives gaseous reaction products, this temperature mainly indicates if carbon or other alloying elements are oxidized. The intensity of the boil can also be estimated in this way. A strong boil forms more carbon monoxide in the exhaust gases at a constant quantity of oxygen added. The carbon monoxide is then burnt in the exhaust-gas hood, creating a higher temperature in the gas flue. A decrease in the speed of rotation slows down the boil, the carbon monoxide is burnt more completely, and the temperature of the exhaust gases is lowered. Consequently, it is easy to estimate by experience the temperature of the exhaust gases that corresponds to a certain combustion of carbon monoxide inside the furnace.

If the production programme is varying and the pig-iron analysis frequently changes, the Kaldo process has the advantage of great flexibility. It is easy to control and modify the heat according to circumstances, as each heat can be treated individually in a similar way to OH practice. Every kind of pig iron can be used and the process can be interrupted at each carbon content desired. Recarbonization is consequently not necessary at higher carbon specifications. Of particular importance is also the possibility of easily adjusting the temperature of the steel to the desired value owing to the great heat surplus produced by the carbon monoxide combustion.

STEEL YIELD

The iron loss in the slag is low in the Kaldo process, primarily because of the high rotation speed which causes the reactions to move closer to equilibrium between steel and slag baths. Particularly in the pre-slag the iron content can be reduced to very low values.

It has proved easy to counteract the formation of foaming slags which always contain a high content of iron also in metallic form. Towards the end of the heat, the iron content of the slag naturally increases with lower carbon content, and for this reason as much of the final slag as possible is kept in the furnace to be utilized during the next heat. In this way it is also possible to make better use of the lime.

As an example, some results can be mentioned which were presented in previous publications.[5, 6] When blowing ordinary basic Bessemer pig iron a first slag, containing only about 5% iron, can be tapped off after oxidation of the major part of the phosphorus content. When blowing silica-rich pig iron with, say, 1%Si, the pre-slagging of silicon can be made practically complete with less than 2%Fe in the pre-slag. An example of this is given in Table II. The iron loss with the first slag is in this case only 0·1%.

The loss of iron with the exhaust gases is also low. During normal operations of basic Bessemer pig iron the loss amounts to between 0·3 and 0·8% of the pig iron according to measurements taken. The steel yield is consequently high in the Kaldo process which is of great economic importance, since each percent by which the yield is improved involves a gain of at least 4*s*/ton of steel.

In Table III the ladle yield of basic Bessemer iron and low-phosphorus iron when cooling with scrap as well as with ore has been put together. The table shows that the steel yield from pig iron + scrap + Fe in the ore is higher when cooling by scrap than by ore.

LINING

Lining life of the Kaldo furnace has not proved quite as good as that of stationary furnaces, although it is gradually improving.

The refractory normally used consists of bricks pressed from tar dolomite, manufactured at Domnarvet. The present type of lining is shown in detail in Fig. 5. With these bricks the lining life has now been improved to 70 heats as against about 50 heats previously when blowing basic Bessemer pig iron, corresponding to a wear of 5·4 mm per heat at the

TABLE IV

Estimated production in two Kaldo furnaces, alternatively in operation, when using basic Bessemer pig iron

Weight of ingots, metric ton/heat		26	100
Time for charging hot metal and additions	min	5	10
blowing	,,	35	55
slag-off, sampling, tapping	,,	30	35
Total time	,,	70	100
Production	tons h	22	60
Tons year (7 500 h)		165 000	450 000

most exposed part of the furnace. When using better pressed and more exactly dimensioned dolomite bricks, it has proved possible to increase the resistance further. With similar bricks from another manufacturer, 79 heats were attained. Tests with 350-mm bricks of burnt magnesite have resulted in a maximum of 90 heats. It turned out, however, that the chemical erosion on these bricks was negligible and that the attack was mainly caused by spalling. New tests are therefore planned with modified bricklaying.

The comparatively short life of the lining depends partly on the relatively long blowing time when blowing basic Bessemer pig iron. Another contributing factor seems to be the high temperature developed inside the furnace by the combustion of carbon monoxide to carbon dioxide. The high speed of rotation does not, on the other hand, seem to affect the lining unfavourably: rather the contrary.

Unfortunately, we have so far no long-term experience of the lining life when using normal low-phosphorus hot metal. It seems, however, to be considerably higher than when using high-phosphorus metal, as the silica pre-slag is tapped off at a relatively low temperature (about 1 500° C). The shorter blowing time in this case is of course also favourable.

As regards the possibility of increasing the lining life in the future, no definite conclusions can be drawn today. A far longer life of the lining must, however, be expected in larger furnaces. The thickness of the working lining can, furthermore, be increased in a large furnace. The shape of the furnace room is also of great importance. In our present furnace, the main attack has been concentrated on the middle of the cylindrical wall. We have been able to overcome this drawback by making the oxygen lance move continuously in a vertical plane. With the movable lance it becomes at the same time easier to regulate the angle between the jet and the bath during the blow and in this way improve the control of the metallurgical reactions.

The total consumption of tar dolomite at Domnarvet amounts to about 25 kg/ton of ingots, of which about 5 kg can be recovered. The total cost of refractories is about 12*s*/ton. This cost is higher than that of stationary converters and to this must be added the drawback of more frequent furnace exchanges. The total difference in cost seems, however, to be more than compensated by the better heat economy of the Kaldo process.

It may be of interest to mention that a considerable number of heats have been carried out in our 3-ton pilot furnace with a lining of silica bricks. The process turned out to be very easy to control and the results were very satisfactory. The lining showed a surprisingly good wear resistance. The attack, which was very evenly distributed, did not amount to more than about 6 kg/ton of steel in this very small furnace, corresponding to a cost of only about 3*s*/ton. Dephosphorization can naturally not be expected when using acid lining, but the modern ore concentrating methods seem to enable a reduction of the phosphorus content in certain ores to such a low value that dephosphorization in the steel process can be omitted.

STEEL QUALITY

Steel quality is essentially a question of good control of the analysis of the steel as well as of the tapping temperature. An essential requirement is that the metalloids, oxygen, phosphorus, nitrogen, and sulphur be kept at sufficiently low values.

The oxygen content of the steel is a key factor, being of great importance for the yield of alloying elements as well as for the control of the solidification process and for securing a clean steel. Investigations have shown that in the Kaldo process the percentage of oxygen is determined by the same [C] × [O] relation as in the OH process with a very good degree of reproducibility.

The upper phosphorus limit that can consistently be kept when blowing a high-phosphorus pig iron varies from 0·015–0·020% in low-carbon rimming steels to 0·030–0·035% in high-carbon killed steels.

The nitrogen content of the Kaldo steel before tapping generally varies between 0·002% and 0·003% even when using an oxygen gas with 95–96% O_2.

The exceptionally low content of sulphur (0·010–0·020%), which is an outstanding feature of the Kaldo steel, is an advantage for the quality in several respects, e.g. good hot workability of steels even with low manganese contents as also with respect to the improved weldability.

A close control of the temperature can be obtained in the Kaldo process according to the aforesaid possibility of an extremely rapid increase in the temperature when necessary. This is of value for making a clean steel with good ingot surfaces. Uniformity with respect to temperature as well as to composition is rapidly secured by means of the rotation which is an additional tool for the quality-control of the Kaldo steel.

The inherent low percentages of phosphorus, nitrogen, and sulphur makes the Kaldo steel especially suitable for purposes where very good cold-forming properties are desired, and several customers have found it superior to the best steels that they could earlier get for such applications.

A comparison has been made between Kaldo steel and OH steel in deep drawing grade with respect to the tensile properties in which a number of heats were rolled into wide strip with thicknesses 0·8–0·9 mm. The rolling, including cold rolling, annealing, and temper rolling, was made in the conventional way under identical conditions.

The average figures obtained for analysis and tensile properties were as follows:

Analysis, %	Kaldo (7 heats)	OH (3 heats)
C	0·074	0·063
Mn	0·34	0·34
P	0·013	0·014
S	0·013	0·021
N	0·002	0·003
Tensile properties		
Yield point, kg/mm²	19·1	20·8
Tensile strength, ,,	32	32·6
Elongation, %	43·7	41·6

In Domnarvet, the Kaldo furnace is mostly used for producing ship-plate steels which have received full approval for all specifications according to, for instance, Lloyd's Register of Shipping. The high impact-ductility, even after strain ageing, and the good welding performance make Kaldo steel very suitable for this purpose. In test heats for different foreign steelworks, a wide range of specifications have been met with carbon contents varying from 0·05% maximum, up to 1·5%, and manganese contents up to 1·5%, covering both rimming, semi-killed, killed, and aluminium-treated steels.

SPECIAL FIELDS OF APPLICATION

Besides steelmaking, it is also possible to use the Kaldo furnace for pre-refining pig iron for OH and electric steel furnaces. A duplex method of that kind will normally turn out more expensive than a direct method, but may under certain conditions be justified. For this purpose the Kaldo furnace has the advantage of securing a good stirring even when refining pig iron with a high silicon content. Furthermore, it is possible to obtain a high steel yield, as the iron content in the slag can be brought down to about 2% before slag-off, as already illustrated in Table II.

The Kaldo furnace can also find use for transforming high-phosphorus iron into hematite cast iron. This problem is of current interest in, for example, the UK, and some experiments have been carried out at Domnarvet in cooperation with BCIRA, the results of which were recently published.[7] As dephosphorization requires a content of iron oxide in the slag that is higher than the content corresponding to the equilibrium with the carbon dissolved in the melt, it is impossible to avoid some decrease of carbon content during dephosphorization. It has, however, proved to be very easy to recarburize the melt with coke after finished dephosphorization and slag-tapping, without too much rephosphorization from the remaining slag.

Suitable agents in the Kaldo process are iron ore or scrap, as mentioned previously. Sponge iron is also under certain conditions a suitable cooling material. It is, of course, not necessary to use a sponge with an especially high degree of reduction. The only drawback will be that a smaller quantity of sponge can be melted per ton of pig iron, if it contains a large amount of iron oxides which has to be reduced. As it is often relatively expensive to produce fully reduced sponge iron, it may be cheaper only partly to reduce the ore in the sponge furnace and charge this product as cooling agent in a Kaldo furnace. An advantage is that the sponge iron can also be charged in fine-grained form.

TABLE V

Estimated production in two Kaldo furnaces when using low-phosphorus pig iron

Weight of ingots, metric tons heat		26	100
Time for charging hot metal and additions	min	5	10
blowing	,,	30	45
slag-off, sampling, tapping	,,	20	25
Total time	,,	55	80
Production	tons h	28	75
Tons/year (7 500 h)		210 000	560 000

If it is desirable to increase the addition of sponge iron, this can simply be attained by a simultaneous well balanced addition of some kind of carbon to the bath. At complete combustion of this carbon to carbon dioxide inside the furnace, the main part of the combustion heat will be utilized in the process. Such an addition of carbon can become economical even when using other cooling agents, such as ore or scrap, to increase the output per ton of pig iron.

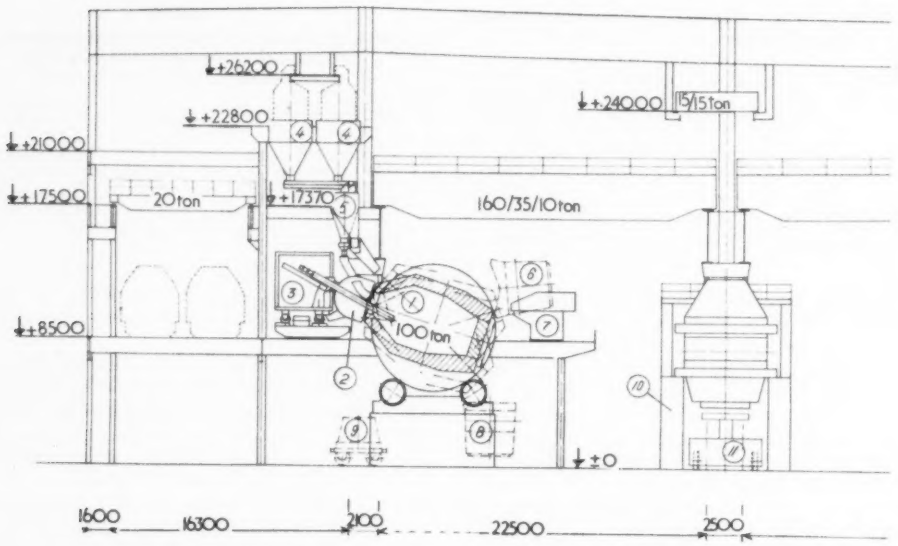
As an example it can be mentioned that when using the Stelling sponge iron method,[8] a sponge iron in powder form can easily be produced with, for instance, the following analysis: 85%Fe (total); 75% degree of reduction; 4·6%C, 3·5% gangue.

Of this product about 600 kg can be melted and reduced per ton of pig iron, containing about 4·0%C, 1%Si, and with low-P content.

These facts encourage the use of the Kaldo process for direct production of steel out of ore, entirely without any addition of pig iron. Instead, the quantity of carbon necessary for the reduction of the ore is charged for instance as coke breeze whereby the carbon monoxide generated is burnt with oxygen to carbon dioxide inside the furnace. A certain quantity of carbon must in this case be added also as fuel. The ore is preferably charged as concentrates. For carrying out the reaction, it is essential that the materials are charged to an iron bath in the furnace, containing a suitable carbon content. After some preliminary tests at Domnarvet there are hopes of creating in this way an economic direct steel-producing method.

PRODUCTION

The plant at Domnarvet has now been in operation for three years, and more than 200 000 tons of ingots have been produced. The capacity was, however, not fully utilized during the first year. The production has now risen to about 18 heats per 24 h. The corresponding tap-to-tap time is about 80 min, of which about 40 min are effective blowing time. By introducing improved charging devices for ore and lime, which are at present under construction, the total time might be brought down to about 70 min, corresponding to 20 charges per day. The annual yield will then be 155 000 tons per year with an operating time of 7 500 h. For larger furnaces, the blowing time will probably be somewhat longer. In Table IV, a comparison is given between a 26-ton and a 100-ton furnace, showing the different time



1 Kaldo converter **2** Fume hood and oxygen lance **3** Hood supporting car **4** Raw-materials storage bins **5** Weighing hoppers **6** Hot-metal ladle **7** Scrap-charging box car **8** Steel ladle **9** Slag pot **10** Relining stand **11** Converter transport car

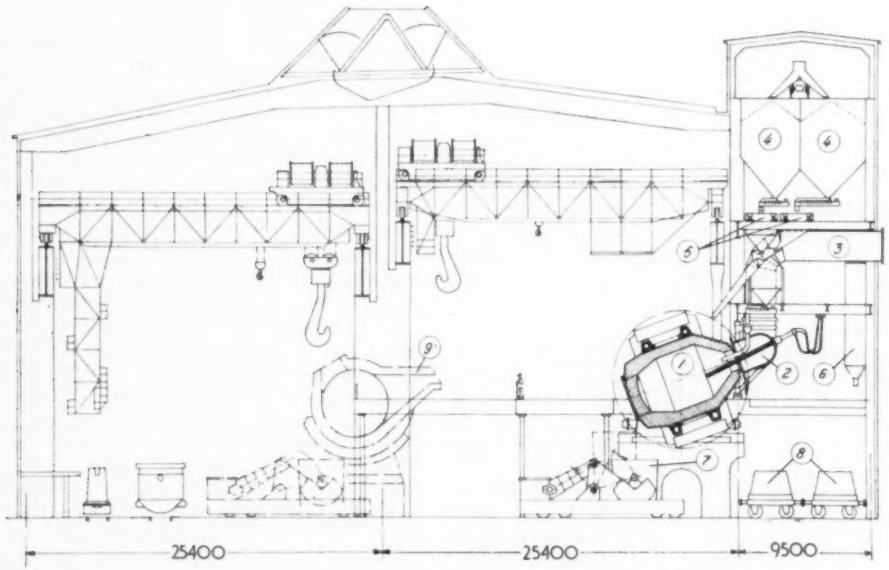
Fig. 6—Cross-section through the Kaldo shop for SOLLAC, France

factors based on our experience, and the annual output.

When refining pig iron with a low phosphorus content it has been found that the effective blowing time can be reduced to about 30 min in the 26-ton furnace, and the time required for slag-taps and sampling is also reduced. The output can therefore be expected to be considerably higher in this case, as is apparent from Table V.

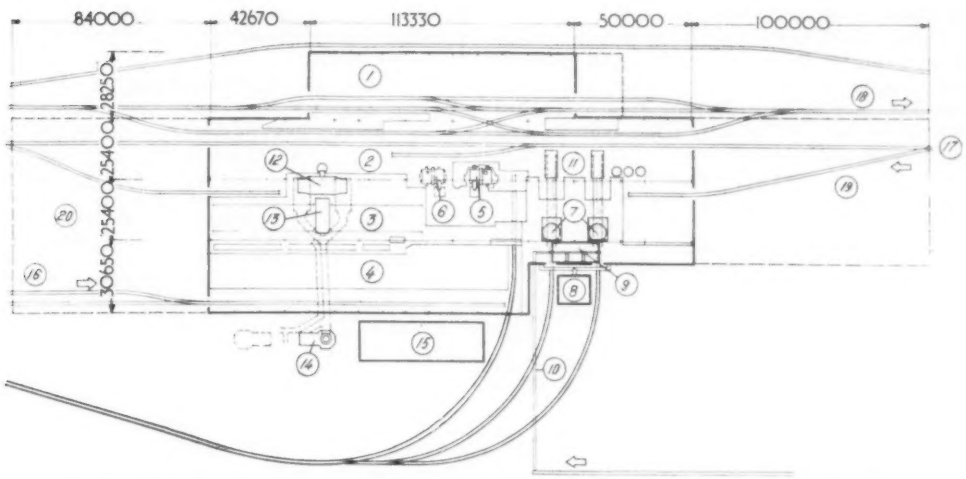
Values in Tables IV and V represent conditions when cooling is done by ore. If cooling is done with scrap, a further increase in output can be expected, the quantity of pig iron per ton steel being considerably lower, and the blowing time consequently shorter. This is, however, to a certain extent counterbalanced by a somewhat longer charging time.

To reach the annual production calculated above, it is necessary to dispose of two furnace bodies, one



1 Kaldo converter **2** Fume hood and oxygen lance **3** Fume duct to cleaning plant **4** Raw-materials storage bins **5** Weighing hoppers **6** Alloy-material bins **7** Steel ladle **8** Slag pot **9** Hot-metal mixer

Fig. 7—Cross-section through Kaldo shop for Oxelösund steelplant, Sweden



1 Mould yard **2** Casting bay **3** Furnace bay **4** Scrap bay **5** 900-ton hot-metal mixer **6** Projected mixer **7** 100-ton Kaldo converter **8** Fume-cleaning plant **9** Raw-materials storage and weighing **10** Raw-materials conveyor **11** 100-ton steel ladle car **12** 150-ton OH furnace **13** 75-ton OH furnace charger **14** Waste-heat boiler and chimney **15** Electrical substation and office **16** Incoming scrap **17** Incoming hot metal **18** Outgoing ingots **19** Future Kaldo plant extension **20** Future OH plant extension

Fig. 8—Plan of steelmelting shop for Oxelösund steelplant, Sweden

of which is operated while the other is being relined. As already mentioned, there are two exchangeable vessels at Domnarvet. A newly relined furnace of this size weighs 125 tons; a heavy crane is therefore required for the exchange of the vessels, also causing increased costs for building and foundations. In most cases it may therefore be more advantageous to install two complete furnaces, each equipped with its rolling stand.

Figure 6 gives an outline of the Kaldo plant under construction at SOLLAC in France, which is expected to start operation in the middle of next year. The capacity of the furnace is 100 tons. Here, the principle of exchangeable furnace vessels has been chosen. The exchange is performed by means of a specially designed transport car on rails. At present only one rolling stand is to be installed. Figures 7 and 8 show the plant which is being built at Oxelösund in Sweden. These furnaces will also have a capacity of 100 tons. In this case, however, two rolling stands, each with one furnace vessel, are being constructed, and the furnaces will be operated alternatively.

It has been proved that the time of operation for a furnace vessel is more than sufficient for stripping and relining the other. It may even be expected that in the future only one extra vessel will be required for the continuous operation of two furnaces.

ECONOMIC ASPECTS

For the reasons stated previously, the production cost for the Kaldo process will be low. Our experience is limited to the plant at Domnarvet but, since it has now been in operation for three years, all costs can be calculated with great safety. The usual basic Bessemer process with oxygen enrichment of the air is used parallel with the Kaldo process at Domnarvet. A cost comparison between the two processes shows that Kaldo steel is at least as cheap as Bessemer steel. This calculation takes into account all costs including those of repair and investment.

As it may be feared that the rapid rotation should cause a considerable wear on the mechanical parts, e.g. the supporting wheels, ball bearings, etc., a thorough investigation was made to discover the extent of wear, and the results indicate that these costs do not exceed 6*d*/ton.

The investment costs for a Kaldo plant might be expected to be considerably higher than for a plant with stationary furnaces, calculated per ton per year, owing to longer blowing time and a shorter lining life. Moreover, a rotating furnace is naturally more expensive to manufacture than a stationary one. This increase in costs is, however, largely counteracted by the considerably smaller gas-cleaning equipment required. Savings can probably also be made in the cost of the building. Moreover, since a higher nitrogen content may be allowed in the oxygen at the same nitrogen content in the steel, the oxygen plant will be somewhat cheaper. The difference in the total investment cost is therefore low and affects the production costs very little. This has also been confirmed by engineering firms which have had the opportunity to make careful comparative studies.

SUMMARY

The Kaldo process has been in commercial use at Domnarvet since May, 1956, mainly for refining basic Bessemer pig iron. The high and variable speed of rotation has proved very valuable for controlling the iron content in the slag, and the intense mechanical stirring causes a desirable intimate contact between slag and metal.

The favourable economy of the process and its reliability have been verified. Advantageous results have been obtained even when refining low-phosphorus pig iron, mainly owing to the particularly good heat economy of the process which allows for large quantities of ore to be reduced into steel, or scrap to be melted during the blow. Outstanding features of the process are, furthermore, low iron losses in slag and

exhaust gases, an effective desulphurization, and very low nitrogen content in the steel. The process is very flexible and can be utilized for refining all kinds of pig iron. Due to a very good control of the process, the specified tapping temperature and final steel analysis can be kept within very narrow limits, and recarburizing is not necessary.

In spite of the comparatively short life of the lining, which, in the last year was improved from 50 to 70 heats, the Kaldo process now produces a cheaper steel than the basic Bessemer process.

ACKNOWLEDGMENT

The authors want to thank The Stora Kopparbergs Bergslags Aktiebolag for their permission to present this paper. We also want to express our gratitude to the management and the steel operation people at the Domnarvet steelplant for their co-operation in developing this process.

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Solute segregation during ingot solidification

By W. A. Tiller, Ph.D.

INTRODUCTION

THE CASTING OF METALS has long been an art with the foundryman. For many years the production of sound ingots has been largely a matter of trial and error, and while a few general principles have been well recognized, the methods used to produce a particular ingot have been mainly the results of the foundryman's hard-won experience.

By experience, he found that an ingot was generally composed of three zones of different structure: (*a*) a small zone of fine randomly oriented grains at the outer wall of the ingot, (*b*) a large columnar zone of preferred orientation crystals extending towards the centre of the ingot, and (*c*) a zone of randomly oriented equiaxed crystals in the centre of the ingot. Within these zones he observed regions in which solute was highly concentrated in the form of a second phase or inclusions and he found that the quality of any ingot was largely governed by the type and degree of this segregation.

At present the forging industry is plagued with the presence of segregation flaws in their large ingots, and it is the purpose of this paper to show that the conventional method of casting has an inescapable size-effect limitation on the quality of ingots. Thus a different method of casting must be employed if sound ingots of large size are required.

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SYNOPSIS

The theory of solidification dealing with the effect of the growth conditions on the mode of solidification is reviewed. The degree of solute segregation is shown to be related to the mode of solidification. The theory is applied to the freezing of an ingot and predicts all the essential features of ingot solidification. It is shown that the formation of oxide and sulphide inclusions depends upon the growth conditions, and that the degree of these defects increases with the size of the ingot. It is also shown that the conventional method of casting must inevitably lead to the formation of deleterious macro-segregation and shrinkage in large ingots. 1528

Although numerous investigations have been made into the subject of segregation, until recently, little attention appears to have been devoted to the macro-structure of the segregated ingot or casting. This is surprising since both segregation and macrostructure are related through the solidification process. In fact the manner or mode of solidification produces certain substructures such as grains, dendrites, lineage structures, etc., and the boundaries of these structures are found to be the regions of segregate. Therefore the segregation that occurs owes its presence to the substructures that are formed as a result of the various modes of solidification.

Following this lead, it is found that the most effective way to study segregation in ingots is to determine the effect of the freezing conditions on the mode of solidification. Thus for various growth conditions this indicates both the type of substructures to expect and the degree of segregation that will accompany them.

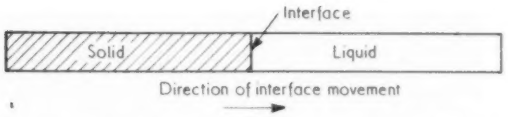


Fig. 1—Unidirectional solidification of binary alloy liquid

To this end, in Part I of this paper the theory of solidification will be discussed to show how the various growth conditions affect the mode of growth of the solid phase. The majority of the material presented in this section has been published elsewhere, and is therefore largely a review of the present state of knowledge in this field. Since this article is mainly directed towards the practical metallurgists involved in the development of alloys and the production of ingots, the previous work pertaining to the verification of the theory is presented for their benefit. It is very likely that they have not had access to this work and are not fully aware of its significance. In Part II, the fundamentals of the theory will be applied to the freezing of an ingot to show how the essential features of ingot solidification can be predicted. The conclusions to be drawn from this section indicate that a new method of casting must be developed if large ingots of high quality are desired.

PART I: THEORY

In the author's opinion all phenomena observed to occur as a result of the solidification of metals can be classified as interactions between three important variables of growth:

(i) Crystallographic features of the growing crystal
(ii) Temperature distribution in the system (solid and liquid)
(iii) Solute distribution in the system (solid and liquid).

The systematizing of the study of solidification into a dependence on these three growth variables leads to a fuller understanding of the prominent features of crystal growth from the melt. The importance of these growth variables in the present study will become apparent from the following treatment.

Consider the solidification of a long cylinder of molten metal from one end as shown in Fig. 1. Let the metal be a binary alloy of concentration C_o

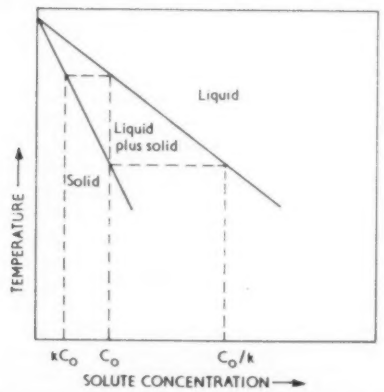


Fig. 2—Portion of binary constitutional diagram for a solute which lowers the freezing point of the solvent

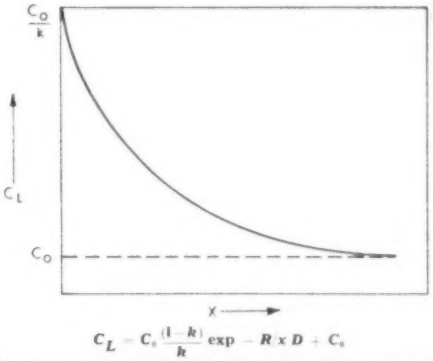


Fig. 3—Steady-state distribution C_L of solute in liquid ahead of the freezing solid liquid interface

and of such a type that the solute lowers the freezing temperature of the alloy as shown in Fig. 2. From the equilibrium diagram, it can be seen that liquid of concentration C_o is in equilibrium with solid of concentration kC_o where k is the ratio of the solubility in the solid to the solubility in the liquid. The k value (distribution coefficient) is a very important parameter in the freezing process and can be of the order of 0·5 as shown here or 10^{-2}, 10^{-3} for some solutes (smaller solid solubility).

As the growth of the solid phase proceeds, there will be a rejection of solute atoms from the solid at the interface. Since the distance through which solute can diffuse in unit time for unit concentration gradient is 2 to 3 orders of magnitude smaller than the rate of advance of the interface R, there will be a build-up of solute in the liquid adjacent to the interface as indicated in Fig. 3. If there is no convection in the melt, the condition will be reached where the amount of solute rejected from the solid per unit time is just equal to the amount of solute diffusing down the concentration gradient at the interface per unit time, i.e. a steady-state solute distribution will exist in the liquid.

This distribution of solute C_L has been calculated by Tiller *et al.*[1] for a plane interface; they found it to be exponential in shape and is represented by equation (1):

$$C_L = C_o\frac{(1-k)}{k}\exp(-Rx/D) + C_o \ldots\ldots (1)$$

where D is the diffusion coefficient of the solute in the liquid, R is the rate of advance of the interface, and x represents distance measured ahead of the interface. The concentration is C_o/k at the interface and falls off as exp $(-Rx/D)$ to C_o some distance ahead of the interface. Thus, if the k value is small, say 10^{-2}, then the concentration at the interface is 100 times the initial concentration. If the solute forms a gaseous phase, this interface concentration may be sufficient to cause the nucleation of a hemispherical bubble of gas on the interface which would lead to the development of a tubular void in the solid as growth proceeds.

An example of this type of gaseous evolution may be found in the solidification of 'rimmed' steel. The presence of carbon and oxygen in solution in the liquid iron will produce the formation of CO gas if

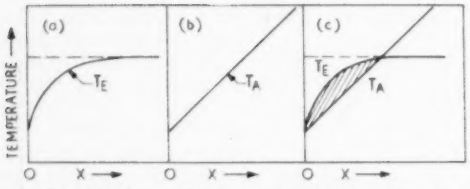


Fig. 4—Equilibrium temperature distribution T_E and actual temperature distribution T_A in liquid ahead of the freezing solid/liquid interface

the concentrations of the C and O are large enough. During the solidification of the ingot, the concentration of both C and O will build up in the liquid ahead of the interface. When these concentrations are large enough, they will react with each other to form CO. The bubbles of CO will be nucleated on the interface as discussed above, and owing to the continued advance of the interface, ' blowhole ' formation will occur.

The thickness of the solute-rich layer is between 10^{-2} and 10^{-4} cm depending on the growth rate. The more rapid is the growth rate, the thinner is this layer. At rapid growth rates, even forced convection is unable effectively to change the solute distribution in this layer.

Since every point in the liquid has a definite concentration of solute, then it also has a definite equilibrium freezing temperature. The equilibrium temperature distribution in the liquid is plotted in Fig. 4*a*. This is the temperature at which the liquid would like to freeze; however, the actual temperature distribution in the liquid is determined largely by heat flow considerations and it may be of the form shown in Fig. 4*b*. If these two temperatures are matched at the interface, then the condition shown in Fig. 4*c* is achieved. The actual temperature at any point in a region is below the equilibrium temperature at the point and this liquid is said to be constitutionally supercooled. When this condition exists, the plane interface will break down into an array of hexagonal cells as shown in Fig. 5. This substructure was studied by Rutter and Chalmers,[2] and they showed that the boundaries are depressions in the interface and are regions rich in solute. This substructure is columnar in the direction of growth, so that a single grain may be thought of as being composed of a bundle of pencils where Fig. 5 represents the ends of the pencils. The axis of the pencil is the major axis of growth, and the pencils are separated from each other by a thin layer of solute-rich material.

The initial point of instability between the plane interface mode of solidification and the cellular interface mode of solidification occurs when the actual temperature gradient is tangent to the equilibrium temperature gradient at the interface as shown in Fig. 6. By equating these two gradients, a borderline condition for the interface instability can be developed:

$$G/R = -mC_o(1-k)/Dk \quad \ldots\ldots\ldots\ldots\ldots\ldots (2)$$

where m is the slope of the liquidus line, and G is the temperature gradient in the liquid at the interface.

From this equation it can be seen that the ratio of the temperature gradient in the liquid G, to the freezing rate R, is proportional to C_o. This then gives

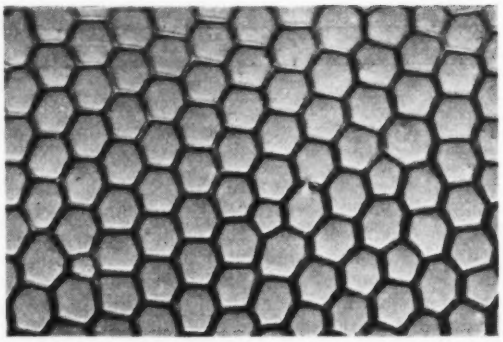


Fig. 5—Photomicrograph of decanted interface showing cellular mode of solidification; Pb–0·1%Sn × 100

an opportunity to test the theory. Equation (2) predicts that if crystals containing C_o of a particular solute (this defines m, D, k) were grown under various ratios of G/R, and the interface examined by rapidly decanting the liquid from the solid, then for a certain range of G/R, a plane interface will be stable and for another range of the ratio, a cellular interface will be stable. Equation (2) predicts the line of demarcation between the regions of stability for the different modes of solidification.

Experiments of this type were carried out by Tiller and Rutter[3] on lead as the base material, and the results for tin as solute in lead are shown in Fig. 7. These show that there is not only a linear relationship between C_o and G/R as predicted, but it agrees quantitatively as well. Figure 8 shows the results for silver as solute in lead, and again there is quantitative agreement with equation (2). From these results it may be seen how little solute it takes to change the mode of solidification. Tiller and Rutter showed that it takes only 5×10^{-6} wt-% Au in Pb to produce the onset of cellular growth with its associated intercellular segregation.

The experiments have completely substantiated the predictions of the theory as to the existence of a

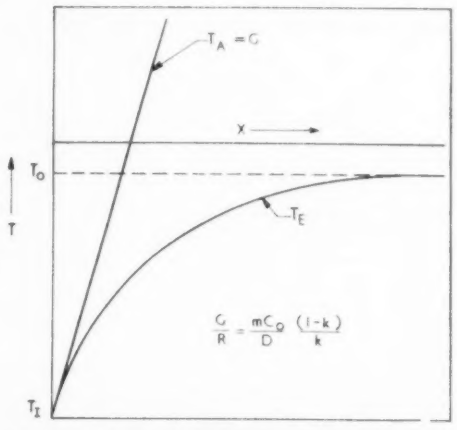


Fig. 6—Growth conditions representing the point of instability between plane interface and cellular interface mode of solidification

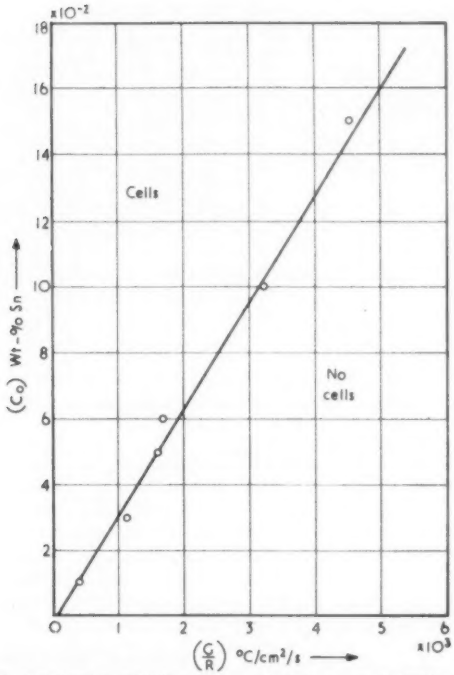


Fig. 7—Critical conditions governing the transition from smooth to cellular interface for Pb–Sn alloys

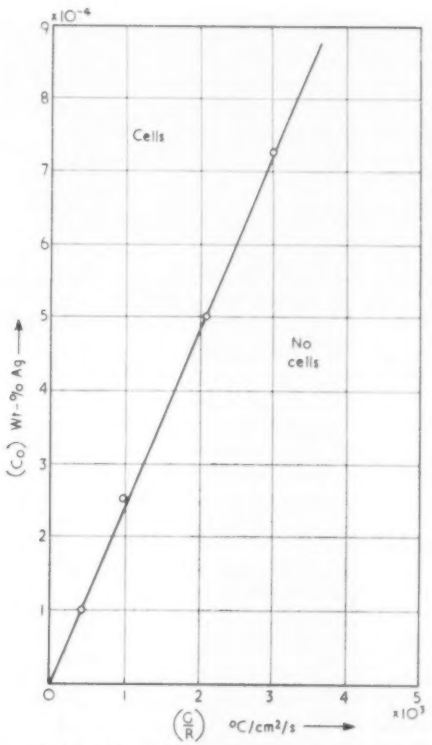


Fig. 8—Critical growth conditions governing the transition from a smooth to a cellular interface for Pb–Ag alloys

solute-rich layer of liquid adjacent to the interface and the quantitative features of this layer. If constitutional supercooling is found to exist under certain conditions of growth, then a plane interface will be unstable and the mode of solidification will be changed to the cellular mode.

Tiller and Rutter showed how the interface appearance changes as the growth conditions are changed from those which produce the plane interface mode of solidification to those which produce the cellular interface mode of solidification. For this, one may lower the temperature gradient in Fig. 6 by increments and study the interface appearance. The results that they found are represented schematically in Fig. 9*a*. Figure 9*b* represents the constitutional supercooling conditions ahead of the interface if it had remained planar, and Fig. 9*c* the probable conditions existing ahead of the interface represented in Fig. 9*a*.

The first evidences of constitutional supercooling were found to be the appearance of very small protuberances called ' pox ' on the interface. The next stage of development is the appearance of irregular cell boundaries and these subdividing into elongated cell boundaries. Finally, the elongated cell boundaries subdivide into more nearly regular cell boundaries. The boundaries form and are stable because they are capable of trapping solute that would normally reside in the interface layer. In this way, they are able to decrease the constitutional supercooling as shown in Fig. 9*c*. Consider Figs. 9*a* and *b*; as one tries to increase the constitutional supercooling in the liquid, the interface decreases it by providing an increasingly greater amount of cell boundary per unit area of the interface in which to store solute.

Figure 10 shows a longitudinal section of the cells and the interface. At the tip of the projection, the growth is in a direction normal to the surface, so that there is some growth in a lateral direction which tends to trap solute between the cells leading to intercellular segregation. The increased solute concentration in the region between two confining cells will decrease the freezing temperature of the cell walls so that they cannot grow together until a lower freezing temperature is reached. However, the tip of the cell is still advancing because the rejected solute from this portion of solid can diffuse away from the interface into the liquid. In this way, the boundary walls may become very deep, narrow grooves during solidification, so that the solid consists of an array of blunt spikes, connected some distance back from the interface, which grow like independent crystals into the liquid. This boundary groove is not observed during decanting as surface tension holds the solute-rich layer of liquid in the groove. It is because of deep boundary grooves of solute-rich liquid that the phenomenon of inverse segregation is sometimes observed.

So that the cells can reduce the constitutional supercooling, they must provide enough boundary to store the solute and they must be of such a shape as to help its diffusion from the tip of the cell to the boundaries in the time allowed by the rate of advance of the interface. In principle, one would expect that the cells could get smaller and smaller to provide the

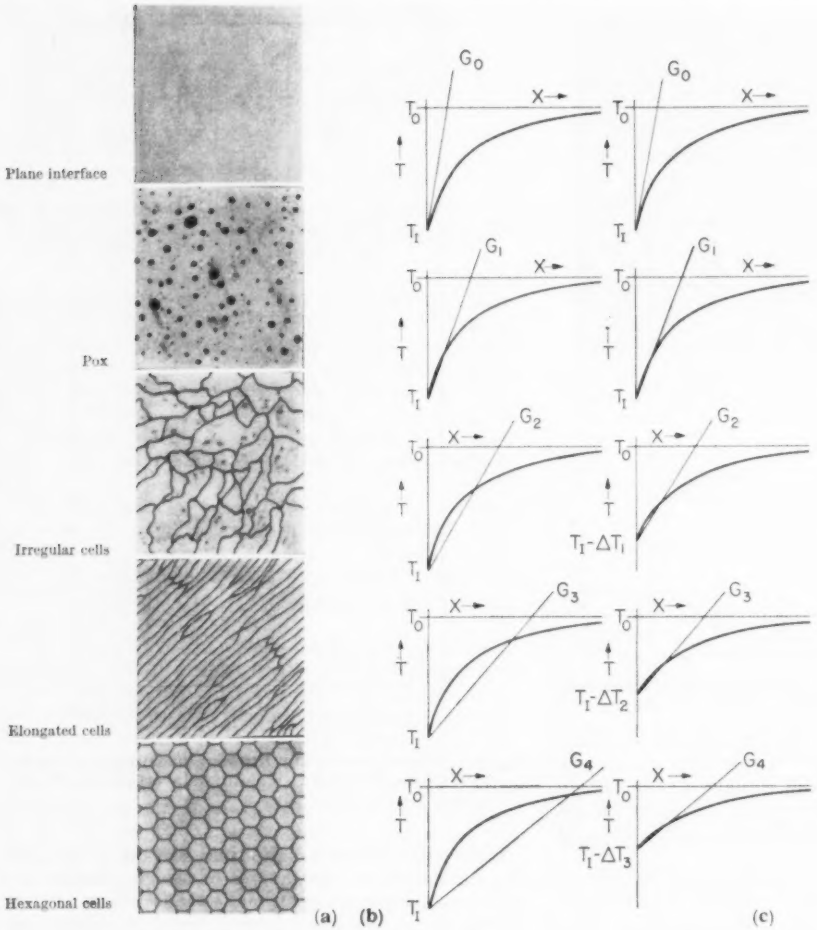


Fig. 9—(*a*) Development of cellular interface (*b*) Constitutional supercooling that would have existed in the liquid if interface had remained planar (*c*) Constitutional supercooling existing in liquid ahead of interface shown in (*a*)

extra boundary. However, this is not possible since the free energy of the system increases as the area of interface increases. There will, therefore, be some lower limit to the size of the cells and some point at which constitutional supercooling is not reduced significantly by the cells because it would require too small a cell size.

There is a limit to the amount of boundary available for storing the solute, but how is the solute able to diffuse to the boundaries. It should be noticed that although the distance through which solute can diffuse in unit time for unit concentration gradient is about 10^{-2} times the rate of freezing, the distance through which heat can diffuse in unit time for unit temperature gradient is 10^2 times the freezing rate. It can be seen, therefore, that the isoconcentrate surfaces in the vicinity of the cells (Fig. 11) will be greatly affected by the interface shape, whereas the isothermal surfaces will be negligibly affected. It can then be assumed that the temperature varies linearly with distance into the liquid, and the temperature on the surface of the cells shown in Fig. 11 must also vary linearly with distance ahead of the base of the cell. The equilibrium diagram in Fig. 2 relates the composition to the freezing temperature in a linear fashion, so that the concentration on the surface of the cell will also vary linearly with distance ahead of the base of the cell.

It can be assumed that the concentration falls off in an exponential manner with distance into the liquid ahead of the cell. From Fig. 12, it can be seen that it is this variation in concentration on the surface of the cell compared to the variation in concentration in the adjacent liquid which produces the lateral

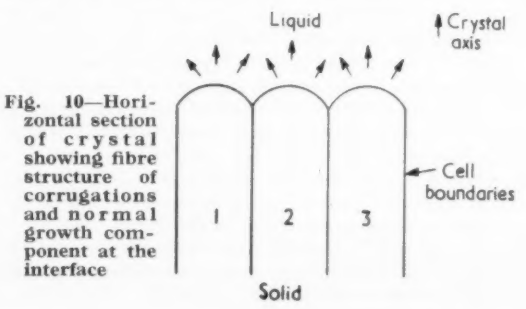


Fig. 10—Horizontal section of crystal showing fibre structure of corrugations and normal growth component at the interface

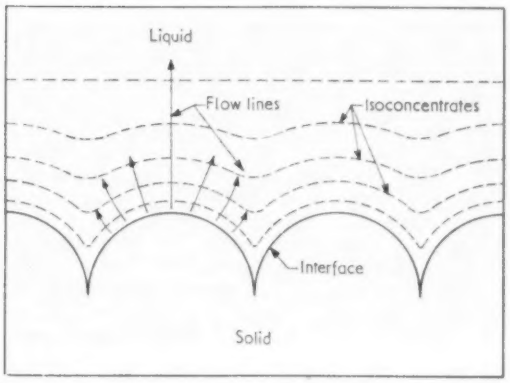


Fig. 11—Isoconcentrate lines in liquid adjacent to a cellular interface

diffusion of solute from the tip of the cell to the region of liquid ahead of the base of the cell. It is the concentration difference between the point c and the point d in Fig. 12 that determines the concentration gradient for the diffusion of solute in this direction. The greater this difference in solute concentration, the greater will be the lateral flow of solute, and thus the greater the possibility of eliminating the constitutional supercooling in the liquid.

However, it has been observed that the farther the cell projects into the liquid (length ad in Fig. 12), the greater the width of the cell becomes (length ae in Fig. 12). This relationship of cell width to cell projection length is caused by the crystallographic mechanism of growth and has been discussed by the author elsewhere.[4] Thus, as the amount of constitutional supercooling to be eliminated is increased, the cells project farther into the liquid. By doing so, they cause the cell to widen, but are still able to enhance the lateral diffusion of solute.

It is found that the cell size ae varies inversely as R and G, and directly as C_o. The k value variation is not known experimentally, but it is probable that the cell size varies inversely as k.

As the cells project farther into the liquid, certain crystallographic effects begin to appear as shown in Fig. 13. The cells on one grain are several times larger

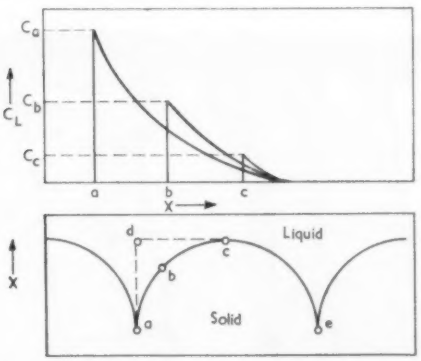


Fig. 12—Solute distribution C_L in liquid ahead of points a, b, and c on the interface. Distributions C_L are cuts through those points parallel to specimen axis

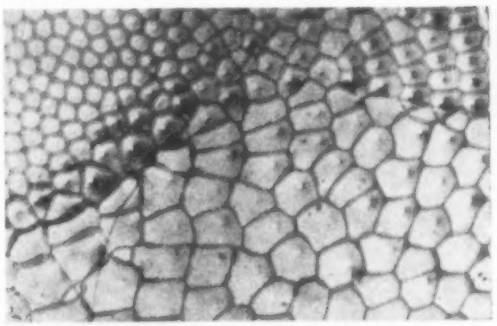


Fig. 13—Interface showing effect of crystallographic orientation on cell size and cell shape; Pb-0·25%Sn × 50

than the cells on an adjacent grain. The grain containing the large cells has a high index orientation interface (low atomic density), the grain containing the small cells has a low index orientation interface (high atomic density). Because of the smaller cell size on interfaces of low index orientation than on high index interfaces, there will be more solute trapped in the boundaries per unit interfacial area. This leads to a lower concentration of solute ahead of this type of interface than a high index one. Thus the low index grain is stable at a higher interface temperature than another grain having a less favourable interface orientation. Because of the interface step that will be produced between two such grains (represented by Fig. 14), the crystallite in advance (small cells), will encroach upon the depressed crystallite and eventually crowd it out of the specimen. It is this difference in cell size on interfaces of different orientation that leads to the <100> preferred orientation in the columnar zone of an ingot of an fcc alloy, and is probably the explanation of the phenomenon in alloys of other crystal structure.

With further increase of the constitutional supercooling, the mode of solidification again changes. The cells begin to develop a branched structure as shown in the Fig. 15. This mode of solidification, although

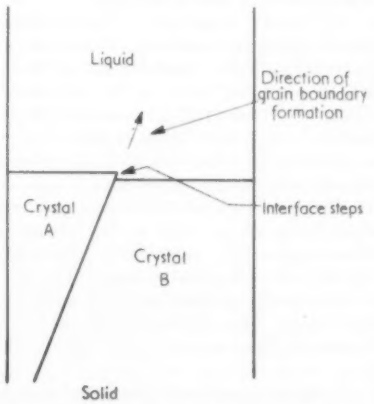


Fig. 14—Bicrystal in which a grain with one interface orientation is freezing at a higher interface temperature than a grain with different interface orientation

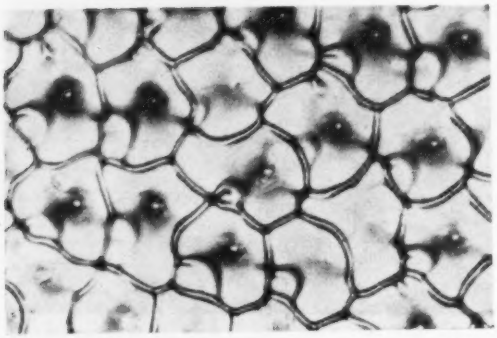


Fig. 15—Interface showing development of side branching on cells; Pb–0·75%Sn × 100

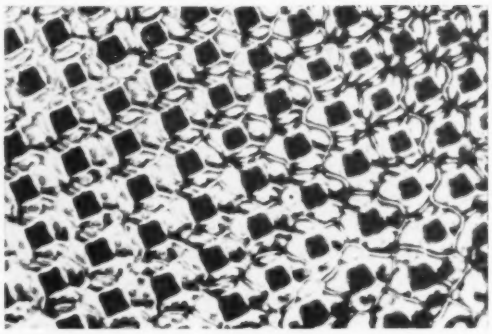


Fig. 16—A (100) interface exhibiting well developed dendrites; Pb–1·0%Sn × 100

called the dendritic mode, is very different from the classical dendrite which grows very rapidly into a supercooled melt and develops a pronounced crystallographic morphology. In the case under discussion, the branched structure is growing in a very controlled fashion under the influence of a positive temperature gradient. Since there is little choice of words to describe this structure, it will be referred to as a dendritic structure merely because it is branched. A dendritic interface for a Pb–Sn alloy (wt-%) is shown in Fig. 16. Figure 17, a photomicrograph of the upper surface of a crystal grown in a horizontal container, shows the branching on the longitudinal cell boundaries.

Figure 18 shows the results obtained for the determination of the growth conditions under which the cellular interface mode of solidification changes to the dendritic mode. Here the plot is of C_o against $(G/R^{1/2})$. From this plot it can be seen that, for a certain range of growth conditions, interfaces of any orientation exhibit the cellular structure. For another range of conditions, some interfaces will exhibit a dendritic structure while low index interfaces will still exhibit the cellular structure. For the final range of conditions, interfaces of all orientations will exhibit the dendritic mode of solidification.

During the latter stages of cell development and the complete range of dendritic growth, the interface is unable completely to eliminate all of the constitutional supercooling, so that as an interface of this type is advancing there will be a zone of liquid ahead of it which is constitutionally supercooled. As the dendrites grow, their size should be governed by the growth conditions in the same manner as the cells, so that the dendrite width might be expected also to vary inversely as R, G, and k, and directly as C_o. When the degree of constitutional supercooling becomes great enough, new crystals will nucleate in the liquid ahead of the interface and halt the growth of the dendrites. This is illustrated in Fig. 19, and has been termed the ' mushy zone ' mode of solidification.

These grains will nucleate and grow radially at some velocity determined by the amount of supercooling present. A solute distribution will be built up at the interface of each grain, and as they grow together, solute will be trapped at the boundaries separating the grains. Tiller *et al.*[1] have shown that in the initial portion of the cylinder of Fig. 1, the concentration in the solid begins at the value of kC_o and increases to the value of C_o in a manner given by equation (3). This equation, although only an approximation, has been shown to be in error by less than 20% by Smith *et al.*[5] From this equation, the length of the initial transient region for various values of k and R is found to be about D/kR. This is the distance in which the concentration in the solid has increased to within 1/e of its steady-state build-up.

$$C_s = C_o\left\{(1-k)\left[1-\exp\left(\frac{-kRx}{D}\right)\right]+k\right\} \quad \ldots\ldots\ldots(3)$$

The characteristic length x of the initial transient region as a function of R for various k_1 is given in Fig. 20. These results will be used in Part II to calculate the rate of build-up of solute ahead of equiaxed grains.

Smith *et al.* have also shown that in the final portion of the cylinder of Fig. 1, the concentration in the solid will increase to an infinite value in an infinitely thin layer at the end of the sample (provided diffusion in the solid is neglected). They calculate that the concentration in the solid will increase from the value C_o, when the interface is greater than a distance D/R from the end of the sample, to a much higher value

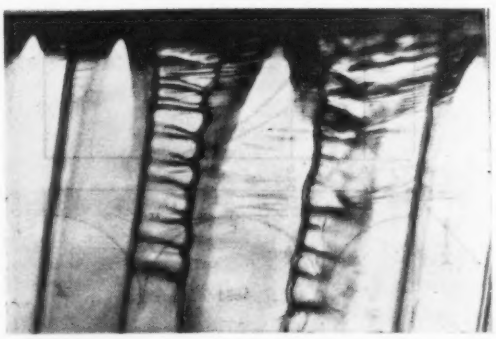


Fig. 17—Upper surface showing initial formation of dendrites as evidenced by oscillation of cell boundary; Pb–0·75%Sn

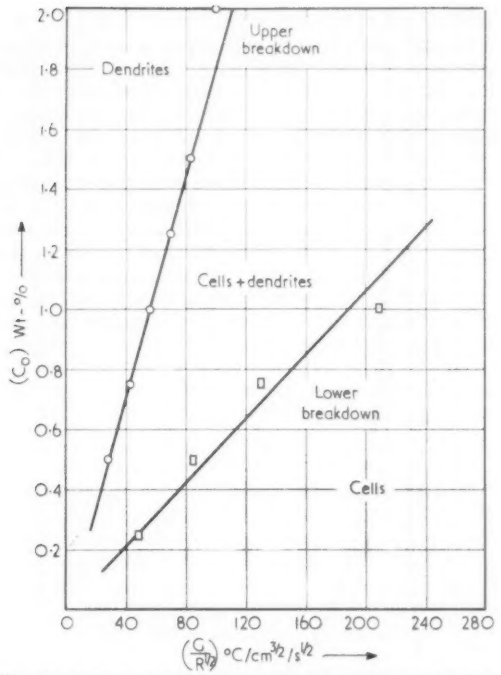


Fig. 18—Effect of growth conditions on transition from cellular to dendritic interface. Upper and lower curves show limits of orientation dependence

as the interface approaches the end of the sample in a manner given by equation (4).

$$\frac{C_s(x_2)}{C_o} = 1 + 3\frac{(1-k)}{(1+k)}e^{-2\left(\frac{R}{D}\right)x_2} + 5\frac{(1-k)(2-k)}{(1+k)2+k)}$$
$$e^{-6\left(\frac{R}{D}\right)x_2} + (2n-1)\frac{(1-k)(2-k)-----(n-k)}{(1+k)(2+k)-----(n+k)}$$
$$e^{-n(n+1)\left(\frac{R}{D}\right)x_2} + \qquad \text{.....(4)}$$

where x_2 is the distance measured from the end of the specimen. Figures 21 and 22 show the variation of the ratio $C_s(x_2)/C_o$ with the dimensionless variable $(R/D)\,x_2$ as a function of various k values. This result will also be used in part II to calculate the amount of solute segregated at boundaries in the equiaxed zone of an ingot. The ratio $C_s(x_2)/C_o$ given by equation (4) is the same as that produced by two grains growing together at the rate R if the solute distributions at the grain interfaces had been given by equation (1) and when the separation distance of the grains $2x_2$ was greater than $2D/R$. This assumes that the grain radius is greater than the thickness of the solute-rich layer in order to neglect edge effects.

PART II: THE FREEZING OF AN INGOT

In the conventional method of casting, metal with a certain degree of superheat is poured to fill a mould. The outer rim of the liquid metal cools rapidly to the melting point and begins to supercool. At some degree of supercooling, nuclei of solid begin to form at random in this supercooled layer at the mould surface. The nucleation of the solid phase occurs by

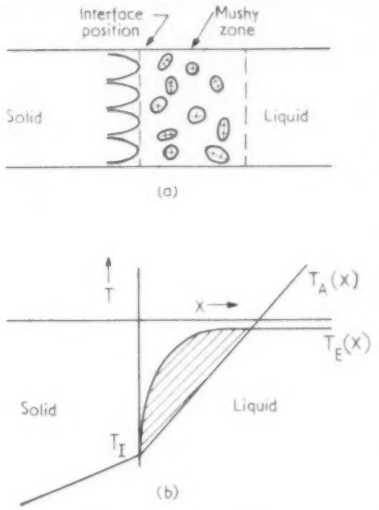


Fig. 19—(*a*) Nucleation of equiaxed grains in liquid adjacent to interface (*b*) Constitutional supercooling existing in liquid ahead of dendrites which is sufficient to cause formation of 'mushy zone'

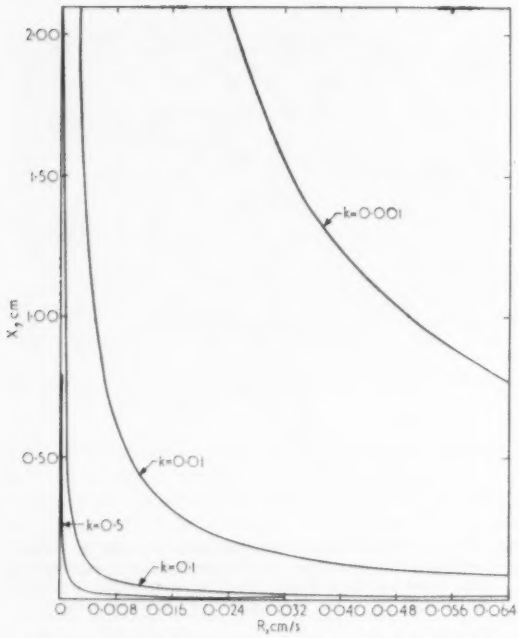
heterogeneous rather than homogeneous means (i.e. by a catalyst); thus, the number of nuclei per unit area of the outer surface will depend upon the number of heterogeneous particles per unit area of the outer surface and how effective they are as nucleating agents. If this number is small, the grains in the chill layer will be large.

The grains in the chill layer are randomly oriented and begin to grow inwards; as they do, two things happen:

(i) Solute begins to build up at the interface, and thus lowers the freezing temperature at the interface

(ii) The temperature gradient in the solid conducts away not only the latent heat of fusion evolved due to growth, but also some of the superheat of the melt so that the temperature gradient in the liquid continues to decrease.

The temperature gradient in the liquid decreases continuously as the interface advances inwards. There will, therefore, be a small initial region where the plane interface mode of solidification will prevail and the solute build-up at the interface will be approximately that calculated by Tiller *et al.*[1] At some later position of the interface, constitutional supercooling must occur and the cellular interface mode of solidification will be stable. Associated with this mode of solidification will be the intercellular segregation. This is the first instance of solute segregation other than that which occurs at grain boundaries during freezing. Since the build-up of solute may not have reached steady state, and since both the rate of growth and the temperature gradient in the liquid are decreasing as the interface advances inwards, the cells will get larger. The cell size will vary with the orientation of the grain and the preferred orientation of the columnar zone will develop. At some still later stage, the interface will again become unstable and



$x = D/kR$, where $D = 5 \times 10^{-6}$ cm²/s, and $k = 0{\cdot}001, 0{\cdot}01, 0{\cdot}1, 0{\cdot}5$

Fig. 20—Effective length of 'initial transient' region x, as function of R for several values of k

will transform into the dendritic mode of solidification. Associated with this new mode will be the interdendritic segregation.

In most alloys, the composition C_o is so great that the plane interface and the cellular interface modes of solidification have an exceedingly short life, and are confined to a distance of a centimetre or less at the outer rim of the ingot. However, they have been discussed in some detail because they elucidate the understanding of solidification.

As mentioned in part I, the dendritic mode of solidification is similar to the cellular mode of solidification except for the appearance of side branches on the cells. The variation in dendrite width with growth conditions should be very similar to the cell width variation; thus the dendrites should become larger as the interface advances inwards since both R and G are decreasing. It can be seen that as the dendrites become larger, less solute can be trapped at the dendrite boundaries if the width of the boundary layer remains constant, and thus the interface concentration must rise. Because of this, the constitutional supercooling in the liquid will tend to increase. At some stage during the growth of the dendritic interface, the temperature gradient will have decreased enough so that part or all of the central portion of the ingot is constitutionally supercooled. When the degree of supercooling is great enough, nuclei of solid will form in this volume of liquid to produce the equiaxed structure.

Before discussing the growth of the grains in the equiaxed zone, the interdendritic channels and their connection with the phenomenon of inverse segregation may be considered.

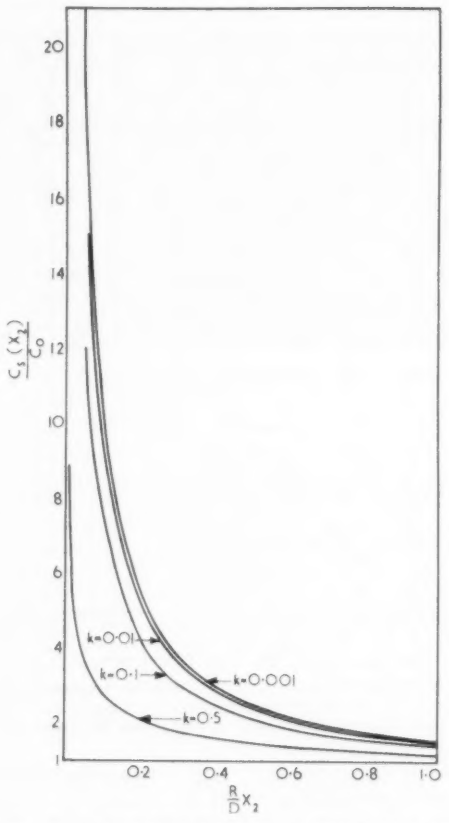


Fig. 21—Terminal transient solute distribution in the solid as a function of the parameter $\frac{R}{D}x_2$

Inverse segregation

Inverse segregation is the reverse of normal segregation. It consists in the segregation towards the cooling face of solute that lowers the melting point of the solvent. The problem has been discussed in some detail by Hanson and Pell-Walpole.[6] They concluded that inverse segregation is caused by interdendritic flow of enriched liquid due to the pressure set up by shrinkage.

It can be seen that the flow of enriched liquid along the interdendritic channels will greatly increase the solute concentration of the regions near the outer surface. In these outer regions, the cell size and dendrite size are small so that there are many channels intersecting unit area of the interface. The cell size increases as the interface advances inwards; thus, the proportion of boundary channel to cell area on the interface decreases as the interface advances toward the centre of the ingot. The freezing of the boundary regions near the outer wall of the ingot draws liquid from the inner channels which, in turn, draws liquid from the interface layer. This flow of material enriches the outermost layers of solid with solute of approximately eutectic composition. The amount of enriched material added to any area parallel to the outer surface will depend upon the amount of

boundary area at that point. Thus, since the dendrite size increases as a function of distance from the outer surface, the amount of enriched material deposited by boundary segregation will decrease as a function of distance from the outer surface. The boundary concentration may be orders of magnitude greater than the dendrite core concentration; thus the boundary segregation may contribute greatly to the average concentration. If it is assumed that the boundary concentration is constant regardless of dendrite size in any ingot, then this will account for the phenomenon of inverse segregation.

Since this type of segregation is due to liquid flow along the channels between the cells and dendrites, the degree of segregation will depend primarily upon the following:

(i) The amount of liquid which must flow along these channels to feed the solid forming at the base of the channels
(ii) The concentration of solute in the liquid freezing at the base of the channel.

The greater the volume change on solidification, ΔV, the greater (i) must be. The longer the channel, the greater will be the average concentration of the liquid at the base of the channel before any freezing begins. This length will depend upon the freezing range of the alloy, and the temperature gradient in the solid since this will determine the distance between the main interface and the base of the channels. The higher the thermal conductivity of the solid, the lower will be the temperature gradient in the solid. Therefore a large ΔV, a large freezing range, and a high thermal conductivity all tend to increase the tendency for inverse segregation. These features are all present in alloys which show the most pronounced examples of inverse segregation, i.e. Al and Cu.

Equiaxed zone

As mentioned previously, when the degree of supercooling ahead of the dendrites is great enough, nuclei of solid will form in this volume of liquid. The number of nuclei forming per unit volume per unit time in this region will again depend on the number of heterogeneous particles per unit volume that will act as nucleating agents. The supercooling at which these grains begin to form will depend on the potency of the heterogeneous particles. As these randomly oriented grains begin to grow from certain centres in the liquid, they will build up a solute distribution ahead of them and, as the interfaces of the grains become closer together, the solute distributions will increase to second-phase proportions in a thin layer as they butt against each other.

In part I, calculations were carried out to show how much of a particular solute could be segregated at these boundaries as a function of the distance x_2 from the separating line between the grains. This calculation assumes that the solute distribution is steady state before the grains begin to butt against each other; in practice, the grain-size may be such that steady state is not reached. For an average freezing rate R, the liquid interface concentration of the grain is given approximately by

$$C_I = \frac{C_0}{k}\{(1-k)(1-e^{-kR^2t/D}) + k\} \quad \ldots\ldots\ldots(5)$$

where t is the time elapsed since nucleation of the grains. If the grain radius is d, then the time required to freeze the grain will be given by $t \approx d/R$. If it is assumed that d is large with respect to the thickness of the segregated layer at the boundary, then the interface concentration as the grains get close together can be given by

$$C_I \approx \frac{C_0}{k}\{(1-k)(1-e^{-kRd/D}) + k\} \quad \ldots\ldots\ldots(6)$$

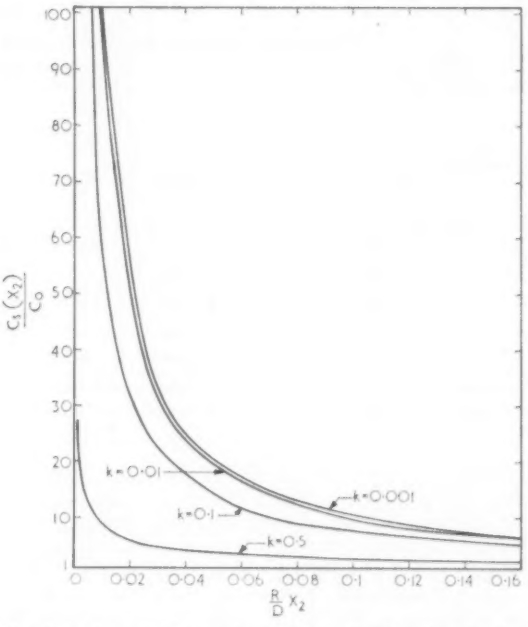


Fig. 22—Terminal transient solute distribution in the solid as a function of parameter $\frac{R}{D}x_2$

When the grains begin to get very close together (10^{-2} to 10^{-3} cm), the presence of the approaching grain will be felt and a build-up of solute will occur between the grains. The calculation for $C_s(x_2)/C_o$ given by equation (4) of part I can be utilized if the solute distribution has reached steady state ahead of the grain. This will be an upper limit in the calculation of the segregation. A lower limit can be had by treating the transient concentration given by equation (6) as a steady state with interface concentration $(\alpha C_o/k)$. If the rate of build-up due to the approach of the other grain is the same as for a steady state interface concentration of C_o/k, then the build-up for C_I will be given by $\alpha C_s(x_2)/C_o$. In fact the build-up should be faster than this, so that the limits on the concentration in the solid for a particular d will be given by $C_d(x_2)$ where

$$\alpha C_s(x_2)/C_o < C_d(x_2)/C_o < C_s(x_2)/C_o \quad \ldots\ldots\ldots(7)$$

The factor α can be calculated from equation (6) since $C_I = \alpha C_o/k$.

Since the theory predicts that a second phase will always be produced provided diffusion in the solid is neglected, even if only in an infinitely thin layer between the grains, then it is the thickness of this layer that will be of interest in the present study. Below a certain layer thickness, the second phase,

TABLE I

Grain-size required for steady-state growth d_{ss} and relative solute segregation $C_s(x_2)/C_0$ at 0·1μ and 1μ from the grain boundary as a function of R and k $D = 5 \times 10^{-5}$ cm²/s

R, mm/min	d_{ss}, cm				$C_s(x_2)/C_0$ at $x_2 = 0{\cdot}1\mu$ and 1μ							
	$k = 0{\cdot}5$	$k = 0{\cdot}1$	$k = 0{\cdot}01$	$k = 0{\cdot}001$	$k = 0{\cdot}5$		$k = 0{\cdot}1$		$k = 0{\cdot}01$		$k = 0{\cdot}001$	
					0·1μ	1μ	0·1μ	1μ	0·1μ	1μ	0·1μ	1μ
1	6×10^{-2}	3×10^{-1}	3	30	50	16	>1000	400	>1000	1000	>1000	>1000
6	10^{-2}	5×10^{-2}	5×10^{-1}	5	18·5	6	1000	32	>1000	45	>1000	55
12	5×10^{-3}	$2{\cdot}5 \times 10^{-2}$	$2{\cdot}5 \times 10^{-1}$	2·5	14	4	300	17·5	800	23	1000	25
24	$2{\cdot}5 \times 10^{-3}$	$1{\cdot}2 \times 10^{-2}$	$1{\cdot}2 \times 10^{-1}$	1·2	10	3	150	9	200	12	300	13
48	$1{\cdot}2 \times 10^{-3}$	6×10^{-3}	6×10^{-2}	6×10^{-1}	6·5	2·5	38	4·5	66	6	72	6·5
240	$2{\cdot}5 \times 10^{-4}$	$1{\cdot}2 \times 10^{-3}$	$1{\cdot}2 \times 10^{-2}$	$1{\cdot}2 \times 10^{-1}$	3	1·3	9	1·6	12·5	1·7	13·5	1·8
1200	5×10^{-4}	$2{\cdot}5 \times 10^{-4}$	$2{\cdot}5 \times 10^{-3}$	$2{\cdot}5 \times 10^{-2}$	1·6	1	2·5	1·1	2·9	1·2	3·0	1·3

if it forms, will not remain stable as a continuous film, but will break up into particles for surface energy reasons. By considering the equations of Smith *et al.*[5] it can be shown that for values of the parameter $Rx_2/D \leq 0{\cdot}1$ the concentration in the liquid is about constant; therefore, if a second phase begins to form at a certain x_2, the width of the second phase region will be $2x_2$. In Table I, the ratio of concentration in the solid to initial concentration as a function of growth rate and k value have been calculated for nominal x_2 values of 0·1 and 1μ. These calculations assume that the grain-size is equal to or larger than the grain-size required for a steady-state build-up. This steady-state grain size d_{ss} is given also in Table I.

From this table, the initial concentration C_o can be 1–3 orders of magnitude below the solid solubility limit and still a continuous film of second phase will be formed at the boundary. Table II gives k values for several elements in δFe. These values were taken from binary diagrams and would, of course, be affected by ternary and quaternary additions. The build-up of oxygen and sulphur at the grain boundaries will be large, so that stable oxides and sulphides will be formed even when the initial concentration of these elements in the melt is very small. The inclusions observed at the grain boundaries of an ingot are, in the author's opinion, formed by the above mechanism and are not mainly due to their presence in the liquid phase before casting.

The formation of FeS at the grain boundaries in Fe is in the form of a continuous film which is deleterious to the mechanical properties of the material. This grain boundary film may be eliminated by the addition to the melt of an appropriate element that produces a more stable sulphide than FeS, i.e. Mn, Ti, or Zr. It is observed that Mn is more effective than either Ti or Zr in eliminating the FeS film from the grain boundary. However, Ti and Zr form more stable sulphides than Mn, and thus one should expect both Ti and Zr to be more effective than Mn. The solution to this apparent paradox comes from considering the distribution coefficients of Mn, Ti, and Zr in Fe (Table II). They are 0·15, 0·6, and 0·5 respectively. From Table I, $R = 1$ mm/min and $x_2 = 1\mu$, gives $C_s(x_2)/C_o = 400$ and 16 for $k = 0{\cdot}1$ and 0·5 respectively. Thus, for the same initial concentration in the melt, about 10 times more Mn will be segregated at the grain boundary than either Ti or Zr. Therefore, even though the Mn forms a less stable sulphide than either Ti or Zr, it will be several times as effective since it segregates preferentially to the grain boundaries on solidification.

A general rule concerned with the replacement of a sulphide film at a grain boundary by a more stable sulphide may be formulated. The addition element that will be most effective is the element which has the smallest distribution coefficient in the particular solvent (e.g. Mg in Ni). This rule can be applied to the replacement of any deleterious grain boundary phase by a more desirable phase.

In this study, interest is basically in a fine grain structure since this gives only microsegregation which can be either annealed out or broken up by working. From the physical principles outlined, qualitative prediction of the factors can be made which will give

TABLE II

Distribution coefficients k for various solutes in Fe, Mo, and Ti

Solute	k in Fe	k in Mo	k in Ti
P	0·2	...	...
Mn	0·15	...	0·3
Mo	0·7	...	...
C	0·25	0·1	~3
O	0·1	...	~6
S	0·002	...	...
Al	0·6	0·3	0·4
Si	0·7	0·3	...
Be	...	0·01	...
B	...	0·02	...
W	...	1·1	2
Fe	...	...	0·3
Ti	0·6	...	...
Zr	0·5	...	0·4

a large ratio of equiaxed structure to columnar structure, and a fine grain structure in these regions.

To ensure a fine grain structure in the chill layer, a large number of effective nucleating particles per unit volume will be needed (i.e. catalysts that will produce solid formation at small supercoolings). This will also lead to a fine structure in the columnar zone. For the same reasons, the nucleating agents will assist in producing a fine grain structure in the equiaxed zone.

To account for the proportion of equiaxed to columnar structure, one must consider Fig. 19 of part I. Here it will be seen that the critical supercooling required for the nucleation of the equiaxed crystals can be affected by varying either the equilibrium temperature distribution or the actual temperature distribution. A small k value or a large C_o will lead to a large equilibrium temperature difference between the interface and the liquid of concentration C_o. This will allow the critical constitutional supercooling required for nucleation to occur at a steep temperature gradient which will exist in the liquid when the interface is close to the outer wall of the ingot. Thus, the greater the freezing range of the alloy, the greater will be the proportion of equiaxed to columnar crystals. By consideration of the actual temperature distribution in the liquid, it can be seen that the smaller the superheat of the melt, the sooner conduction through the mould wall can lower the temperature gradient to the critical value of constitutional supercooling. This will occur closer to the outer wall of the ingot.

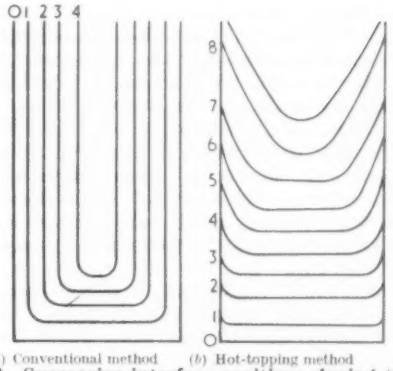
The work of Northcott[7, 8] substantiates these predictions. His experimental results can be accounted for on the basis of this theory, but it requires an analytical treatment to deal with them in detail and this will not be considered here.

The foregoing describes qualitatively the essential features of ingot solidification. Analytical treatments of the various aspects discussed here are under way and details will be published later.

Ingot size effect

The foregoing treatment is straightforward for ingots of reasonable size where the temperature gradients are sufficient to control the solidification under all the various modes of solidification. However, when the freezing of large ingots is considered, it will be seen that there is no escape from large units of macrosegregation.

In a large ingot cast in the conventional fashion, after a thick wall of solid metal has formed, the temperature gradient in the solid has decreased to a very small value and the temperature gradient in the liquid has decreased to almost a zero value. Therefore, when grains begin to nucleate and grow in the equiaxed region, their latent heat will remain in the vicinity of this region for an appreciable period of time. This will inhibit the nucleation of more equiaxed grains and the first few crystals to nucleate will grow large. Because the conduction of heat away from this region will occur slowly, the grains will grow slowly. There will therefore be large regions of solute segregation when these grains grow together and butt against the columnar zone. Since the rate of freezing



(*a*) Conventional method (*b*) Hot-topping method

Fig. 23—Successive interface positions during the conventional method of ingot solidification as compared to a hot-topping or continuous deposition method

of these equiaxed grains is an order of magnitude smaller for large ingots (100 in. dia.) than for small ingots (10 in. dia.), from Table I one can expect that the thickness of the second phase region may be an order of magnitude larger in large ingots than in small ones for the same C_o. Conversely, for a large ingot the C_o values for oxygen and sulphur will need to be 1 or 2 orders of magnitude smaller than for a small ingot if the same degree of oxide and sulphide formation is not to be exceeded.

Perhaps even more disastrous to the properties of the ingot is the fact that accompanying this macrosegregation there will be a macroscopic shrinkage effect as large volumes of trapped liquid freeze. This shrinkage will occur in the regions last to freeze which also are the regions of greatest segregation. Thus there is a second phase plus one of the following:

(i) High stress concentration
(ii) Porosity
(iii) Shrinkage cavity.

This size effect inherent in the conventional method of casting seriously limits the quality of large ingots and a different method of casting must be devised which will not have this size effect limitation if sound ingots of large size are desired. Any such method must be able to satisfy three basic constraints. First, it is necessary to eliminate any shrinkage effects on a macroscopic scale and minimize them on a microscopic scale. The method of casting must, therefore, be such that liquid is able to feed the central portion of the ingot. A method utilizing a controlled deposition rate of liquid metal is needed for this. Figure 23 represents the relative interface positions as a function of time in ingots cast by conventional practice and by a hot-topping or continuous deposition method. In one ingot, a pronounced shrinkage cavity will form, in the other, no shrinkage cavity will form. The important point to consider here is the shape of the interface. As the diameter of the ingot is scaled up the deposition rate must be adjusted accordingly to maintain a desirable interface shape. The second constraint is concerned with the inhibition of the nucleation of equiaxed crystals, or at least the control of their size. To do this, it is necessary to eliminate any large degree of constitutional supercooling or at

least confine it to a narrow region ahead of the interface. This can be done either by maintaining a steep temperature gradient in the liquid or by eliminating the solute distribution by forced convection in the melt. In practice, a combination of the two may be attained. The third constraint involves the lowering of the C_o level of those impurities that create inclusions. Fortunately, these constituents are either gases or can combine with other elements to form gaseous products which can be extracted from the melt by maintaining a vacuum in contact with the molten metal.

Two methods of casting which satisfy these three basic constraints and are therefore capable of eliminating the macroscopic defects from ingots are: (*a*) consumable electrode vacuum-arc melting,[9] and (*b*) drip-pool melting.[10, 11] The vacuum-arc melting is in a more advanced stage of development than the drip-pool melting techniques, and thus will probably be the method first utilized for the production of large ingots.

CONCLUSIONS

The theory of solidification, in its present stage of development, is capable of predicting the essential features of ingot solidification. Some of these features are:

(i) Rimming action in steel to produce blowhole formation
(ii) Development of a preferred orientation in columnar grains
(iii) Segregation to substructure boundaries
(iv) Inverse segregation by interdendritic flow of enriched liquid
(v) Ratio of equiaxed zone to columnar zone dependence on superheat and freezing range
(vi) Formation of inclusions at the boundaries of equiaxed grains. The theory, which appears to have been well substantiated by experimental fact, can now be exploited to enable the foundryman to devise methods for increasing the quality of ingots.

By applying the theory to the solidification of large ingots, it has been shown that severe macrosegregation and shrinkage are unavoidable if the ingots are produced by conventional methods; however, theoretically it is possible to produce ingots of any size without macrosegregation or shrinkage if the proper solidification method is used.

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Boron-treated cast steels

By W. J. Jackson, B.S., M.Sc.(Eng.), A.R.I.C., A.I.M.

COMPREHENSIVE REVIEWS of the development of boron steels have appeared previously[1, 2, 3] and need not be discussed in detail. However, brief mention will be made of the more important aspects of the effects of boron.

It is generally accepted that the most outstanding effect of the addition of boron to low-alloy steel is the consequent increase in hardenability.[4–7] This effect decreases with increasing carbon content, and is virtually negligible in the eutectoid composition.[8, 9] It also rapidly decreases with increase of nitrogen content in solid solution.[9, 10] Other factors which influence the improvement of hardenability are austenitizing temperature and rate of cooling. At

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SYNOPSIS

In the first part of this investigation, boron additions were made to low-carbon molybdenum cast steels. Reproducible tensile, impact, magnetic, and depth properties were obtained. In the normalized condition, consistently high tensile strengths and good magnetic properties were exhibited. The impact properties were not high but were of the same order as may be expected from carbon steels at a similar tensile level. For castings where high impact strengths are not required, the low-carbon molybdenum–boron steels appear to have a useful application.

In the second part of the investigation, boron additions were made to 1½%Mn, 1½%Mn–Mo, 1¼%Cr–Mo, and 1½%Ni–Cr–Mo cast steels. By the Jominy end-quench test, it was found that the hardenability of each steel was increased, particularly that of the 1½%Mn–Mo steel. In fully hardened sections tempered to equal hardness, tensile properties were not significantly affected by the boron additions. In fully and incompletely hardened sections, impact properties of some boron-treated steels were somewhat better, yet for others slightly worse than those of the boron-free steels of otherwise similar composition. It was concluded that improvement in impact properties was not necessarily associated with increased hardenability resulting from boron treatment, and that the replacement of conventional alloys by boron could not be recommended.
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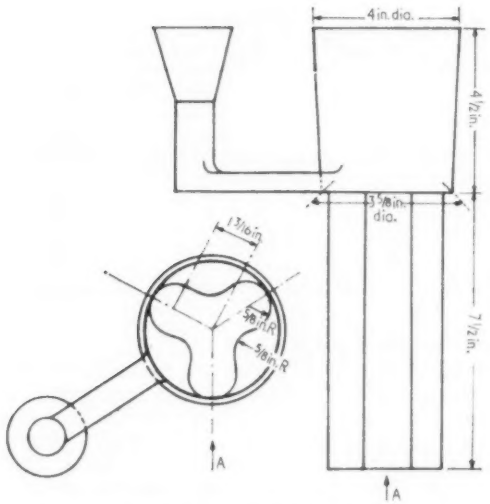

Fig. 1—Clover-leaf test block casting

high austenitizing temperatures, the effect on hardenability is decreased,[5, 8] but can be recovered by lowering to a normal austenitizing temperature before quenching.[7] The effect of boron does not appear when the rate of cooling is comparatively slow.[8, 9]

The amount of boron required in the steel to improve hardenability is very small, probably less than 0·01% being as effective as a few percent of the more conventional elements.[3] The optimum quantity is variously quoted as 0·002%,[4] 0·001%,[5] 0·002–0·005%,[9, 11] and 0·003–0·008%.[12] The addition of boron should, however, be preceded by strong de-oxidizers and de-nitrifiers, and be made to the ladle, otherwise the boron will be lost or occur in an ineffective form.[9]

Published data on mechanical properties obtainable in boron steels are less plentiful than that on hardenability, and relate mostly to wrought steels. At equal hardness levels boron is reported to have no effect on tensile and yield strengths of fully hardened and tempered steels.[6, 13] The effect on impact strength appears to be largely dependent on the hardened section size and the tempering treatment, but boron can markedly lower the impact values.[6] Slightly lower tempering temperatures are required for equivalent strength and hardness for boron-treated steels than for other steels of equivalent hardenability.[4, 13]

From the foregoing, it follows that low-alloy boron-treated steels should be quenched and tempered to secure best mechanical properties associated with improved hardenability. There is one important exception, however, in that a low-carbon molybdenum–boron steel* has been found to show markedly superior tensile properties in the normalized and even in the as-rolled conditions. The development of this high-strength steel, which also exhibits very good weldability, has been described by Bardgett and Reeve.[14] Although this steel does not require quenching, the cooling rate in normalizing must exceed a critical rate which is called the 'limiting rate of cooling' (L.R.C.), for the improved tensile properties to be realized. The austenite transformation then takes place at a lower temperature. The effect is somewhat analogous to full hardening, but should not be confused with it, since the cooling rate never approaches that necessary for the formation of martensite.

It is clear, therefore, that there are two distinct applications of boron in steels:

(i) To low-carbon molybdenum steels, to improve tensile properties in the normalized condition
(ii) To medium-carbon low-alloy steels, to improve hardenability.

Published work concerning these applications relates almost exclusively to wrought steels. One notable exception is the work of Gertsman, Faurschou, and Buhr[15] who investigated the mechanical properties of low-carbon molybdenum steels treated with boron and rare-earth metals. The mechanical properties of cast medium-carbon boron steels have been reported on,[16] while the use of boron to increase the wear resistance of cast steels has also been described.[11, 17] High-boron (4–6%) cast steels for application in atomic energy engineering have been investigated[18] but the scope of these steels is entirely different from that at present under discussion, and no further reference to them will be made.

* This steel is the subject of British Patent No. 611598 and is registered as 'Fortiweld.'

TABLE I

Results of chemical analyses of low-carbon molybdenum–boron steels

Heat No.	Condition	Composition, %															
		C	Si	Mn	S	P	Ni	Cr	Mo	Ti	Al*	Boron			Nitrogen		
												Acid Sol.	Acid Insol.	Total	Sol.	Insol.	Total
1	As-cast	0·107	0·33	0·61	0·036	0·039	0·10	0·13	0·49	0·013	0·24	0·0007	0·0023	0·0030	0·0020	0·0045	0·0065
	Normalized	...	...	...	...	...	...	...	...	...	...	0·0011	...	...	0·0020	0·0045	0·0065
2	As-cast	0·114	0·32	0·75	0·040	0·042	0·10	0·07	0·58	0·010	0·21	0·0017	0·0033	0·0050	0·0025	0·0040	0·0065
	Normalized	...	...	...	...	...	...	...	...	...	...	0·0018	...	...	0·0020	0·0045	0·0065
3	As-cast	0·109	0·33	0·77	0·033	0·039	0·11	0·09	0·49	0·011	0·27	0·0027	0·0029	0·0056	0·0020	0·0035	0·0055
	Normalized	...	...	...	...	...	...	...	...	...	...	0·0027	...	...	0·0010	0·0040	0·0050
4	As-cast	0·115	0·31	0·71	0·039	0·036	0·10	0·10	0·46	0·012	0·24	0·0010	0·0016	0·0026	0·0025	0·0030	0·0055
	Normalized	...	...	...	...	...	...	...	...	...	...	0·0011	...	...	0·0010	0·0045	0·0055

* Spectrographic determination

TABLE II

Values of LRC and lower and upper critical points of low-carbon molybdenum–boron heats

Heat No.	L.R.C. Value (° C/min)	Ac_1, ° C	Ac_3, ° C
1	50	745	970
2	50	745	960
3	45	750	975
4	50	745	960

Because of the meagre information available on the mechanical properties of boron-treated cast steels, it was decided to carry out an investigation with the following aims in view:

(i) To establish the extent to which the tensile, impact, and magnetic properties of low-carbon molybdenum–boron cast steels† are reproducible. Also to investigate the influence of a double normalizing as compared with a single normalizing treatment.

(ii) To establish for a series of medium-carbon low-alloy steels of varying total alloy content, the extent to which hardenability may be increased by boron treatment. Also to investigate the variation in mechanical properties in the fully hardened and tempered condition, and the variation in properties owing to ' mass effect.'

SECTION I: LOW-CARBON MOLYBDENUM–BORON STEELS

Provision of material

Steels were made in a 100-lb h.f. furnace. Details of the steelmaking procedure are given in Appendix I. Each heat of steel was cast into four moulds to give clover-leaf test block castings of the dimensions shown in Fig. 1. Chemical analyses of each of the four heats are given in Table I and nitrogen and soluble-boron contents are quoted for both the as-cast and normalized conditions.

As an additional check on the effectiveness of the boron addition, the L.R.C. test was carried out on each heat before heat treatment of the test material. This test involves dilatometric determination of the slowest cooling rate to give a ' lower ferrite ' transformation, which is associated with the improvements in tensile properties.[19] The L.R.C. values are given in Table II, together with the Ac_1 and Ac_3 temperatures, determined from the dilatometer heating curves.

† ' Fortiweld ' in cast form.

TABLE III

Comparison of cooling rates[19]

L.R.C. value, ° C/min	Approx. equivalent, in.	
	Plate thickness	Bar diameter
90	1	2
50	1½	3
38	2	4
25	3	6

Heat treatment and testing

Feeder heads and test leaves were removed from all the test block castings by sawing. Individual test leaves of about 1¼ in. dia. were used for heat treatment, for despite the rather low acid soluble (effective) boron content of heats nos. 1 and 4, the L.R.C. values of 45–50° C/min indicated that improved properties should be maintained in bars up to about 3 in. dia., or in parallel sided sections up to about 1½ in. thickness (*see* Table III).

Single normalizing treatments were given by heating the test leaves to 1 000° C, holding at temperature for 1 h, and cooling in air. This procedure was repeated for the double normalizing treatments. Higher austenitizing temperatures were required for this type of steel because of the combined effects of low carbon, and about 0·5%Mo in raising the Ac_3 temperature to about 970° C (*see* Table II).

One standard 0·564 in. dia. tensile test piece was machined from each heat in each of the single and double normalized conditions, and tested in a 50-ton universal testing machine. Twelve Charpy 10 × 10 × 60 mm V-notch impact test pieces were also machined from each heat in each condition of heat treatment. Impact tests were made on a standard Charpy machine, over a temperature range of − 40° to + 130° C. The method of attaining the required testing temperature is given in Appendix II.

Hardness tests on representative specimens and test-leaf sections showed that there was no variation along the length of the test blocks or transversely through the test leaves. The overall hardness was found to be very consistent.

For magnetic permeability tests on this material, bars of the following dimensions were machined from each heat in the single normalized condition:

TABLE IV

Tensile properties of low-carbon molybdenum–boron steels in the single and double normalized conditions of heat treatment

Heat Treatment	Heat No.	U.T.S., tons in²	Proof Stress, tons in²				Limit of Proportionality, tons in²	Elongation on $4\sqrt{A}$, %	Reduction of Area, %
			0·02%	0·05%	0·1%	0·2%			
Single Normalized	1	44·3	27·4	29·5	30·5	32·0	15·0	24·0	49·1
	2	46·5	27·3	28·7	29·3	30·7	15·3	23·5	47·1
	3	45·2	24·7	25·7	27·1	28·7	12·0	23·0	50·4
	4	44·9	24·7	26·5	27·7	29·1	12·0	24·0	48·6
Double Normalized	1	45·7	27·7	30·9	31·3	32·6	14·0	23·0	48·4
	2	46·8	27·5	28·9	29·9	30·9	13·0	23·0	45·5
	3	45·3	27·0	27·9	28·7	29·8	16·5	23·0	51·2
	4	44·7	26·5	27·9	28·5	29·9	14·5	24·0	49·9

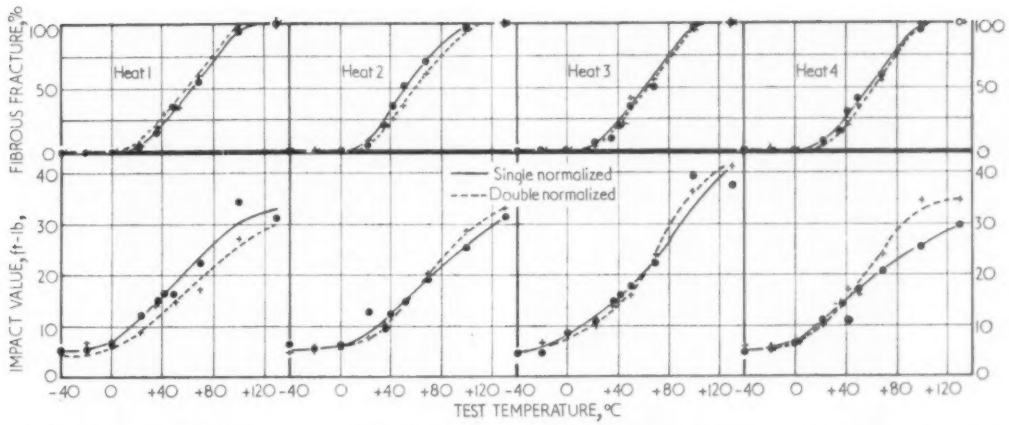

Fig. 2—Charpy V-notch impact energy and fracture appearance transition curves for four heats of low-carbon molybdenum–boron cast steel

Width	2·7 cm
Thickness	0·4 cm
Length	17·5 cm (approx.)

A secondary coil of 50 turns of 36 s.w.g. enamelled wire was wound on to each bar, over an insulating layer of American tape. A permeameter of 5 in. effective length was used to obtain the B/H readings.

Test results and discussion

Tensile tests—Tensile test results, including proof stress measurements are given in Table IV, which shows that in both the single and double normalized conditions of heat treatment, consistently high tensile properties were obtained. It was clear that double normalizing offered no advantages in so far as improved tensile properties were concerned.

Impact transition tests—Impact energy and fracture appearance transition curves were drawn for each heat in each condition of heat treatment. The curves are shown in Fig. 2. Transition temperatures, using the criteria of mean impact energy and 50% fibrous fracture, were interpolated from the curves, and these results are given in Table V.

The results show that the impact properties are fairly consistent, and while not as good as low-alloy cast steels heat treated to similar tensile levels, they are equally as good as carbon steels made to B.S. 592 grade C, or B.S. 1760 grade A. In the latter specifications, impact tests are not called for but if carried out, the impact values would probably not exceed 10 ft-lb, which is the average room temperature Charpy V-notch value obtained in these experiments. In spite of the slight grain refinement resulting from the double normalizing treatment (*see* Fig. 3) the impact properties were not improved.

Magnetic permeability tests—B/H curves were drawn for magnetizing field strengths up to 80 Oe, as shown in Fig. 4. It is apparent that the magnetic properties of this material, in the range tested, are inferior to the requirements of B.S. 1617: 1957* grade *A* or *B* material. However, the magnetic properties appear to be reproducible, and the ductility of this material (*see* Table IV) is sufficient to meet the requirements of B.S. 1617: 1957, which specifies a minimum elongation of 22% for grade *A* material. In general, the magnetic properties were favourable for material of such a high tensile level.

TABLE V

Mean energy and 50% fibrous fracture transition temperatures of low-carbon molybdenum–boron steels in two conditions of heat-treatment

As determined by the V-notch Charpy impact tests

Heat Treatment	Heat No.	Mean Hardness V.P.N., 50 kg load	Mean Energy Transition Temp., °C	50% Fibrous Fracture Transition Temp., °C
Single Normalized	1	226	+53	+65
	2	232	+65	+50
	3	229	+62	+62
	4	226	+50	+58
Double Normalized	1	231	+68	+62
	2	237	+65	+62
	3	230	+65	+65
	4	228	+56	+62

Conclusions

The following conclusions are based upon tests carried out on four consecutive heats of low-carbon molybdenum–boron steel, cast into clover-leaf moulds from a 100-lb basic lined h.f. furnace, the addition of boron by ferro-boron being preceded by thorough de-oxidation with aluminium and ferro-titanium.

(1) The tensile properties obtained in test bars normalized in 1¼-in. dia. section were consistently good, 45 tons/in² U.T.S., being associated with an elongation of 23%, a reduction of area of 50%, and a 0·1% proof-stress/U.T.S. ratio of 0·65.

(2) The impact properties were low in comparison with low-alloy steels heat treated to similar tensile levels, but were equally as good as the higher-tensile carbon steels which are not normally tested for impact strength. The average room temperature Charpy

* B.S. 1617: 1957: 'Mild steel castings of high magnetic permeability.'

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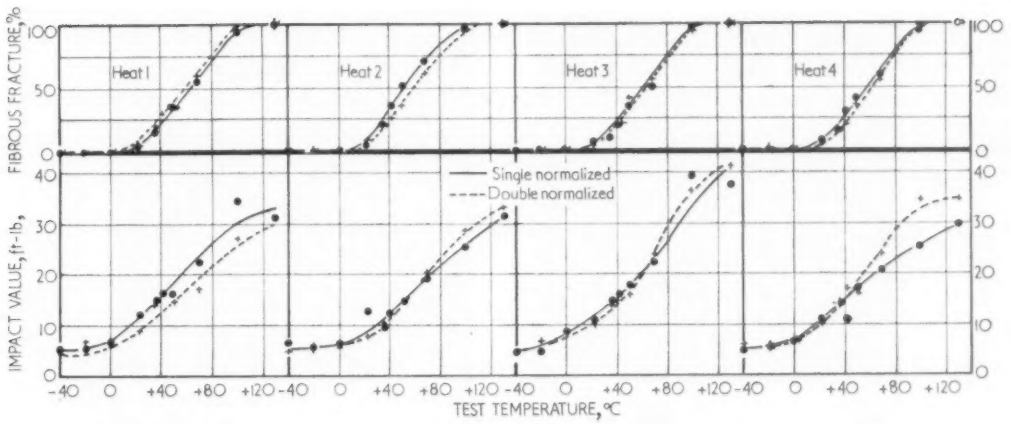

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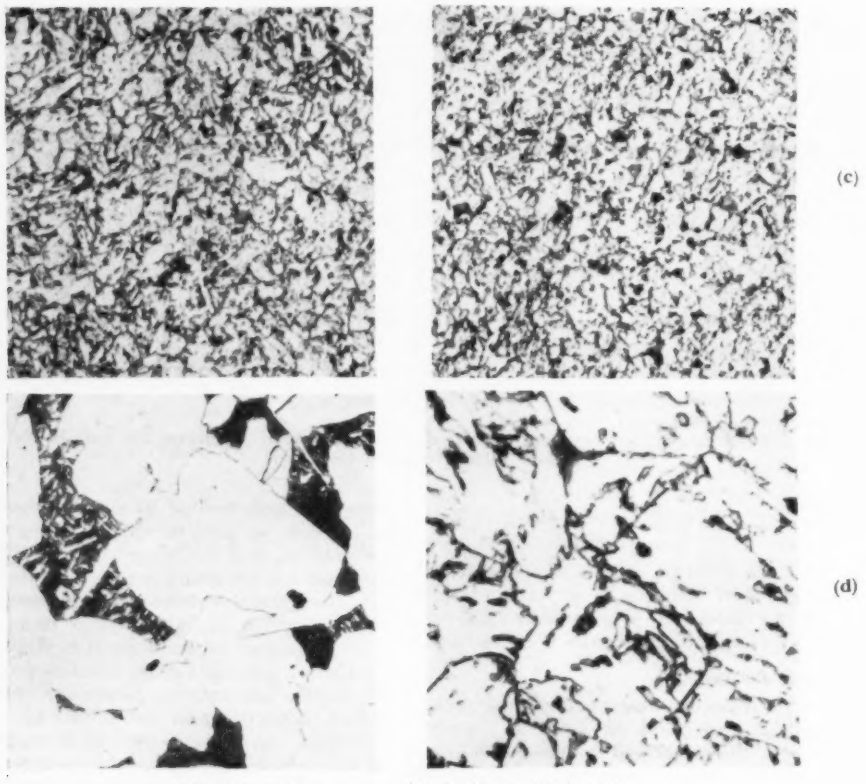


(*a*) As-cast, × 200
(*b*) Single normalized, × 200
(*c*) Double normalized, × 200
(*d*) Single normalized, × 1 000

Fig. 3—Microstructures of low-carbon molybdenum–boron cast steel, heat no. 1, etched 2% nital

V-notch value was of the order of 10 ft-lb. The 50% fibrous fracture impact transition temperatures were in the range 50–65° C.

(3) The L.R.C. test indicated that the improved tensile properties associated with this steel can be consistently developed in the normalized condition, in sections up to 3 in. dia. (bars) or 1½ in. thickness (plates).

(4) No advantage was gained by double normalizing low-carbon molybdenum–boron steels.

(5) Magnetic properties were found to be reproducible. At field strengths of less than 80 Oe however, magnetic properties were inferior in comparison to the requirements of B.S. 1617: 1957 grade *A* or *B* material.

SECTION II: LOW-ALLOY BORON STEELS

Provision of material

Steels were made in the same h.f. furnace as was used for making the low-carbon molybdenum–boron steels. Steelmaking procedure was similar, and details are given in Appendix I. Four boron-free and four boron-treated heats were made in the following types of steel:

1½%Mn, 1½%Mn–Mo, 1¼%Cr–Mo, 1½%Ni–Cr–Mo.

Four clover-leaf test block castings, of the dimensions given in Fig. 1, were cast from each heat, the analyses of which are given in Table VI.

The general purpose of the investigation has been stated previously. The testing programme fell naturally into three parts:

(i) Jominy end-quench tests, to determine the extent to which hardenability is increased by the addition of boron
(ii) Tensile and impact transition tests to determine the effect of boron on low-alloy steels in the fully hardened and tempered condition
(iii) Impact tests on incompletely hardened material of different section sizes, to determine the influence of boron on ' mass effect.'

Heat treatment and testing

Feeder heads were removed from all the test-block castings, which were then annealed for 3 h at 950° C. Individual test leaves were then removed by sawing.

One Jominy end-quench test piece was machined from each heat, in accordance with specification ASTM 255–48T. All test pieces were heated to and end-quenched from 900° C, with the exception of the 1½%Ni–Cr–Mo test pieces, for which a temperature of 870° C was used. Hardness traverses were made on flats ground on the test pieces after cooling.

To prepare material in the fully hardened and tempered condition, individual leaves from each heat were hardened by oil quenching and tempered to a nominal 50 tons/in² tensile level (equivalent hardness 230 B.H.N.). Details of these heat treatments and

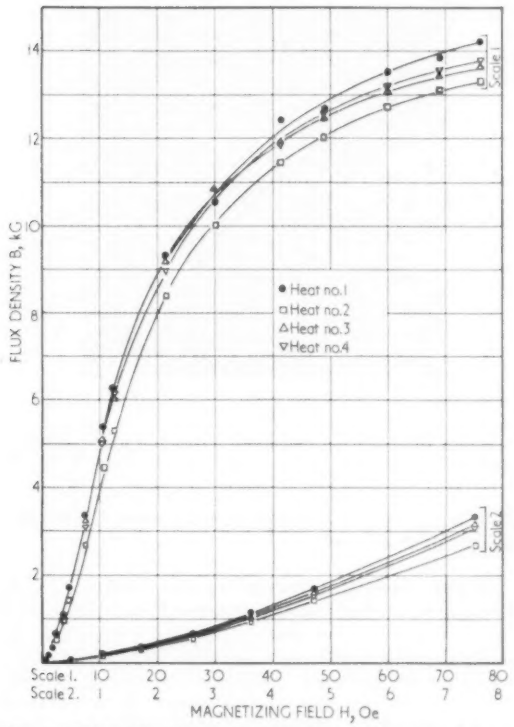

FLUX DENSITY B, kG
MAGNETIZING FIELD H, Oe
Scale 1. 10 20 30 40 50 60 70 80
Scale 2. 1 2 3 4 5 6 7 8
Heat no.1
Heat no.2
Heat no.3
Heat no.4
Scale 1
Scale 2

Fig. 4—Magnetization curves for four heats of low-carbon molybdenum–boron steels in single normalized condition of heat treatment

the hardnesses obtained are given in Table VII. The 1½%Mn steels were water quenched from the tempering temperature to minimize temper embrittlement. A decreased resistance to tempering of all the boron containing steels except the 1½%Mn–Mo may be observed. The microstructure of the 1½%Mn–Mo boron-treated steel indicated that martensitic hardening was more complete than in the absence of boron (*see* Fig. 5). Microstructures of the other type of steels appeared to be unaffected by the addition of boron.

TABLE VII

Heat treatment* applied to low-alloy boron-treated and boron-free steels in single-leaf section (about 1½ in. dia.)

Heat No.	Steel Type	Harden, h °C	Temper, h °C	Hardness Attained, B.H.N.
5	1½%Mn	2 900 O.Q.	3 565 W.Q.	220, 225
9	1½%Mn + B	2 900 ,,	2 565 ,,	229, 229
6	1½%Mn–Mo	2 900 ,,	1 660 A.C.	229, 217
10	1½%Mn–Mo + B	2 900 ,,	2 690 ,,	231, 231
7	1½%Cr–Mo	2 900 ,,	3 690 ,,	230, 231
11	1½%Cr–Mo + B	2 900 ,,	2 690 ,,	230, 230
8	1½%Ni–Cr–Mo	2 900 ,,	3 690 ,,	228, 228
12	1½%Ni–Cr–Mo + B	2 900 ,,	2½ 690 ,,	235, 232

* Test blocks annealed 3 h 950° C before sectioning

One standard 0·564 in. dia. tensile test piece was machined from each test block and tested in a 50-ton universal testing machine. Twelve 10 mm × 10 mm × 60 mm Charpy V-notch impact test pieces were also machined from each test block. Hardness tests on representative specimens indicated that there was no large variation along the length, or transversely across the test leaves. Impact tests were made on a standard Charpy machine over a temperature range − 155° to + 135° C. The method of attaining the required test temperature is given in Appendix II.

For determination of the 'mass effect' with the material available, it was decided to simulate oil hardening of larger sections by air hardening smaller specimens machined to sizes calculated to give cooling rates equivalent to the larger sections when oil hardened. The method used was that reported by Steven and Mayer,[20] who deduced that the equivalence relationship was satisfactorily represented by the equation:

$$\log D_A = 1{\cdot}59 \log D_o + \log b$$

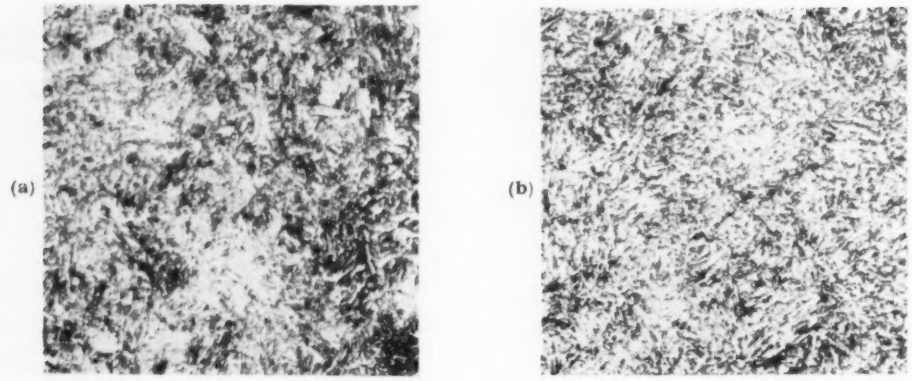
where D_A is the diameter (in inches) of the bar which must be cooled in air to simulate the oil-cooling of a bar of diameter D_o, and b is a constant, in this case having a value of 0·052 for deriving the equivalent air-cooling rate at the centre of an oil-quenched bar.

Four round specimens for air cooling were machined

TABLE VI

Results of analyses of low-alloy heats, with and without boron treatment

Heat No.	Steel Type	Composition, %										
		C	Si	Mn	S	P	Ni	Cr	Mo	Boron		
										Acid Sol.	Acid Insol.	Total
5	1½%Mn	0·26	0·39	1·62	0·038	0·043	0·10	0·08	0·01	...	...	...
9	1½%Mn + B	0·24	0·33	1·55	0·037	0·045	0·10	0·06	0·03	0·0019	0·0023	0·0042
6	1½%Mn–Mo	0·26	0·37	1·58	0·038	0·047	0·10	0·05	0·27	...	...	...
10	1½%Mn–Mo + B	0·27	0·41	1·60	0·037	0·046	0·09	0·11	0·32	0·0017	0·0028	0·0045
7	1½%Cr–Mo	0·29	0·32	0·63	0·037	0·040	0·12	1·34	0·34	...	...	...
11	1½%Cr–Mo + B	0·25	0·38	0·73	0·037	0·040	0·14	1·37	0·39	0·0022	0·0034	0·0056
8	1½%Ni–Cr–Mo	0·29	0·31	0·58	0·034	0·033	1·47	1·22	0·40	...	...	...
12	1½%Ni–Cr–Mo + B	0·27	0·32	0·62	0·035	0·038	1·49	1·20	0·36	0·0002	0·0012	0·0014



(a) (b)

(*a*) 1½%Mn–Mo (*b*) 1½%Mn–Mo + B

Fig. 5—Microstructures of 1½%Mn–Mo cast steels with and without the addition of boron, hardened and tempered, etched 2% nital × 200

from each test block, to simulate the oil quenching of 3, 4, 5, and 6-in. dia. bars. The length of each specimen was at least five times its diameter, to minimize the effect of end cooling. Heat treatment was essentially the same as that given to the individual clover leaves (*see* Table VII). The two smaller sized specimens, however, were in each case held for 1½ h at 900° C instead of 2 h. For air cooling, the specimens were suspended in mid-air and shielded from extraneous draughts.

Subsidiary standard 5 mm × 5 mm × 30 mm V-notch impact test pieces were machined from the round bars and were tested at room temperature (+ 20° C) on an Avery 120 ft-lb capacity Izod machine which had been modified for testing beam-type test pieces.

Test results and discussion

Hardenability tests—The Jominy curves shown in Figs. 6*a–d* indicate that although the hardenability of all the steels has been increased, the most outstanding increase resulting from the boron addition has occurred in the 1½%Mn–Mo steel. The hardenability of the 1½%Ni–Cr–Mo steel is seen to be improved by as little as 0·0002% acid-soluble boron. This result was unexpected, it previously being considered unlikely that boron would substantially increase the hardenability of this steel.

Tensile tests—Tensile test results, including proof stress measurements are given in Table VIII. It will be seen that in the case of the 1½%Mn steel, the addition of 0·0019% acid-soluble boron apparently improved the ductility. This, however, may be partly accounted for by the slightly lower carbon content of heat no. 9 (0·24%C) compared with heat no. 5 (0·26%C). The 1½%Mn–Mo steel, the hardenability of which was increased to a marked extent by the boron treatment, appeared to show no significant change in ductility owing to the addition of 0·0017%B, although lower values of proof stress were exhibited. Exogenous inclusions were observed, however, in the fracture of the test piece from heat no. 6; therefore, in all probability the ductility figures should be higher than those quoted.

The addition of 0·0022%B to the 1¼%Cr–Mo steel appeared to lower tensile ductility, but the carbon content of heat no. 7 is higher than heat no. 11 (0·29% compared with 0·25%); therefore the difference in ductility may not be significant. In the case of the 1½%Ni–Cr–Mo steel, the addition of 0·0002% acid-soluble boron appeared to increase the

TABLE VIII

Tensile-test results of low-alloy boron-treated and boron-free steels

Heat No.	Steel Type	U.T.S., tons in²	Proof Stress, tons in²				Elongation on 4√A %	Reduction of Area, %
			0·02%	0·05%	0·10%	0·20%		
5	1½%Mn	50·0	11·0	15·0	22·2	33·0	15·0	23·0
9	1½%Mn + B	50·8	22·4	25·9	32·8	42·0*	19·5	41·9
6	1½%Mn–Mo	46·5	28·4	30·8	34·0	35·2	21·0	38·5
10	1½%Mn–Mo + B	46·4	14·2	18·0	24·0	38·0	21·0	33·5
7	1¼%Cr–Mo	47·5	32·2	35·0	36·6	37·2	26·0	48·4
11	1¼%Cr–Mo + B	48·2	24·3	27·9	34·7	38·8	20·0	44·8
8	1½%Ni–Cr–Mo	51·3	29·8	30·2	31·2	33·2	10·5	15·9
12	1½%Ni–Cr–Mo + B	49·9	37·0	38·2	38·8	39·8	18·0	36·6

* Extrapolated

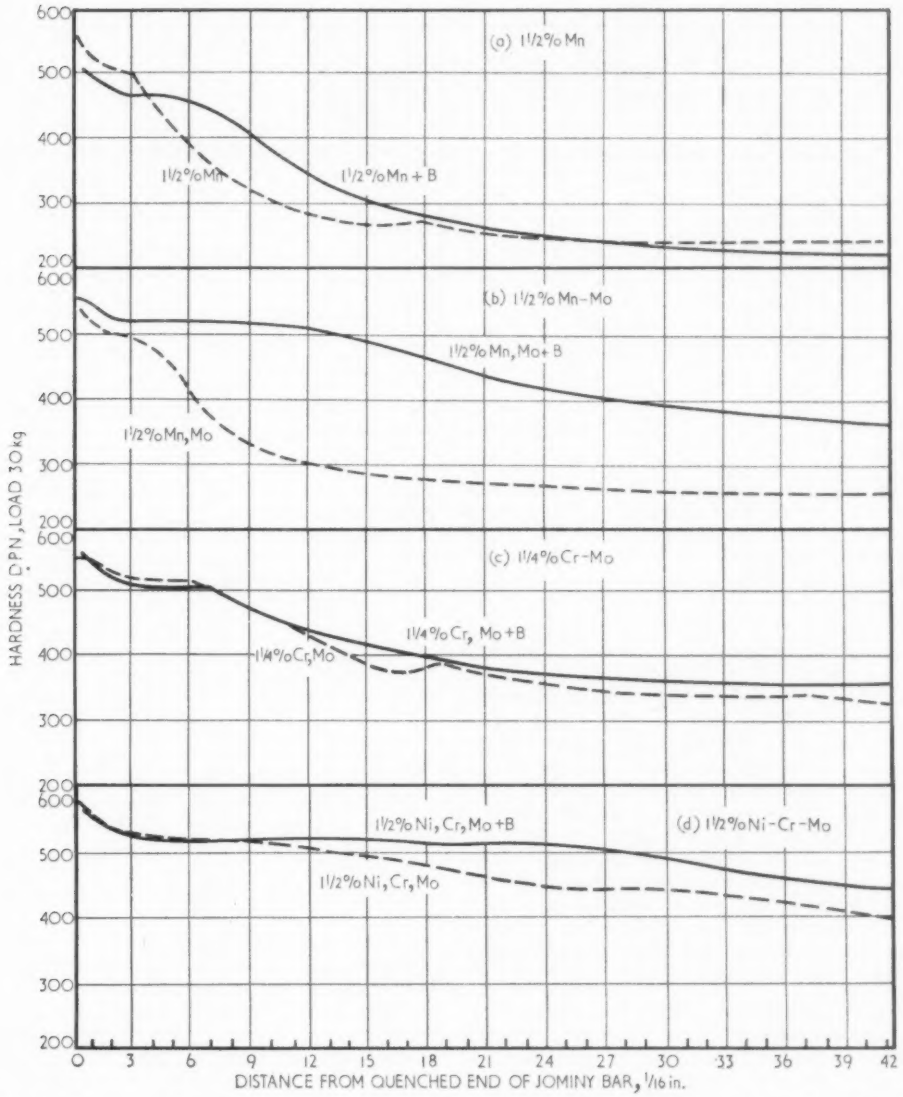


Fig. 6—Jominy end-quench hardenability curves for steels as shown, with and without boron

ductility. It is pointed out, however, that a micro-examination of the test piece from heat no. 8 revealed a chain of inclusions, which would constitute a plane of weakness and account for a lowered ductility.

The general conclusion to be drawn from the tensile test results is that the presence of 0·0014–0·0056% total boron in the steels tested did not significantly alter tensile ductility.

Impact transition tests—Impact energy and fracture appearance transition curves were drawn for each material, and are shown in Fig. 7. Transition temperatures, using for criteria the temperatures to give 15 ft-lb impact energy, mean impact energy, and 50% fibrous fracture, were interpolated from the curves and are given in Table IX.

The scatter of impact values obtained in impact testing was small with the exception of heat no. 8 (1½%Ni–Cr–Mo steel) which gave scattered results. Slight differences in mean hardness between the boron-treated and boron-free heats being compared were considered insignificant in view of the comparatively large transition shifts obtained.

Examination of the results shows that, with the exception of the 1½%Mn steel, the addition of boron lowered the maximum impact value at the higher testing temperatures (i.e. when fracture is 100% ductile). The effect of boron on the transition temperature appeared to depend on the compositional type of steel to which it was added. This opinion is also held by other investigators,[2, 21, 22] and it appears

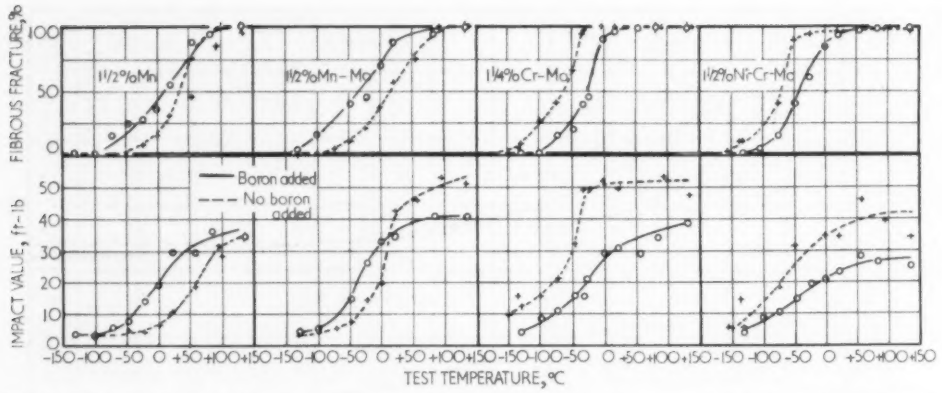


Fig. 7—Charpy V-notch impact energy and fracture appearance transition curves for low-alloy steels with and without boron

that the effect of boron can vary for a particular steel type within the compositional limits of its specification.

In the case of the 1½%Mn and 1½%Mn–Mo steels, the transition temperatures were lowered, whereas with the 1¼%Cr–Mo and 1½%Ni–Cr–Mo steels, the transition temperatures were raised. However, the impact transition properties of the 1½%Mn and 1½%Mn–Mo steels without boron were inferior to those which may be expected from these steels,[23] and on the basis of one heat in each case, it could not be concluded with certainty that the improvements shown by the boron additions were significant.

It is pointed out that other investigators[9, 21] have shown that the effect of boron on impact strength is strongly dependent on tempering temperature, although there is no agreement on the temperatures which give a reversal of the boron effect. Furthermore, it has been reported[24] that boron exerts a direct influence in increasing susceptibility to temper brittleness, when tempering is carried out in the range 425–565° C.

'Mass effect' impact tests—Room-temperature impact and hardness values made on subsidiary standard test pieces are given in Table X. For comparison of results given by standard and sub-standard test pieces, one sub-standard test piece was also machined from each of the heats nos. 8–11. Results are shown graphically in Fig. 8.

Using published hardness data[25] in conjunction with the Jominy end-quench curves (Figs. 6*a–d*), the maximum fully hardened round-bar diameters were found to be:

Elemen	*Dia., in.*	*Element*	*Dia., in.*
1½%Mn	1·45	1½%Mn + B	2·25
1½%Mn–Mo	1·55	1½%Mn–Mo + B	>4·75
1¼%Cr–Mo	3·8	1¼%Cr–Mo + B	>4·75
1½%Ni–Cr–Mo	>4·75	1½%Ni–Cr–Mo + B	>4·75

(i.e. full hardening to 50% martensite, assuming $H = 0{\cdot}5$ for oil quenching). Using these results with the impact results shown in Fig. 7, it is clear that for the 1½%Mn–Mo and 1¼%Cr–Mo types, there are round-bar sizes where the boron-free steels are incompletely hardened but where the boron-treated steels are fully hardened. In both types of steel, the impact strengths of the incompletely hardened specimens were superior to the fully hardened boron-treated specimens. It must be concluded, therefore, that improvement in impact strength is not always associated with increased hardenability.

In the case of the 1½%Mn steels, neither of which harden fully in sections greater than 2·25 in. dia., the boron-treated steel gave better impact properties in the unhardened specimens. Thus it would appear

TABLE IX

Transition results of low-alloy boron-treated and boron-free steels Based on three arbitrary criteria, as determined by V-notch Charpy impact test

Heat No.	Steel Type	Mean Hardness V.P.N., Load 30 kg	15 ft-lb: Temperature, °C	Mean Energy Transition Temp., °C	50% Fibrous Fracture Transition Temp., °C
5	1½%Mn	267	+36	+50	+40
9	1½%Mn + B	258	−20	− 5	+17
6	1½%Mn–Mo	241	−22	− 4	+16
10	1½%Mn–Mo + B	220	−45	−25	−34
7	1¼%Cr–Mo	233	−117	−66*	−61
11	1¼%Cr–Mo + B	234	−42	−20	−23
8	1½%Ni–Cr–Mo	253	−86	−65	−64
12	1½%Ni–Cr–Mo + B	244	−48	−40	−37

* Curve extrapolated beyond −150° C

TABLE X

Impact results and hardness of subsidiary standard V-notch Charpy specimens Machined from oil-quench simulation bars of low-alloy boron-treated and boron-free steels

Heat No.	Steel Type	Impact Value,* ft-lb					Hardness V.P.N., 30 kg load				
		Leaf dia., 1¼ in.	Equivalent oil-quenched bar diameter				Leaf dia., 1¼ in.	Equivalent oil-quenched bar diameter			
			3 in.	4 in.	5 in.	6 in.		3 in.	4 in.	5 in.	6 in.
5	1½%Mn	...	7½	5	4½	5	...	231	228	217	223
9	1½%Mn + B	6	6½	6	7	6	266	234	217	201	201
6	1½%Mn–Mo	...	7	7	7	6	...	221	218	214	210
10	1½%Mn–Mo + B	5	5½	5	4	4	233	235	223	229	218
7	1½%Cr–Mo	...	7½	7	6	7½	...	249	233	234	224
11	1½%Cr–Mo + B	5	6	...	6	5½	241	246	252	237	238
8	1½%Ni–Cr–Mo	5	7½	7	7½	8	250	250	245	238	234
12	1½%Ni–Cr–Mo + B	...	5	5½	4	4	...	258	269	249	248

* Subsidiary standard V-notch beam type specimen, tested on 120 ft-lb machine

that boron treatment can improve mechanical properties without improving hardenability.

For the 1½%Ni–Cr–Mo steels, the maximum round-bar diameters in which full hardening to 99·0% martensite occurs (again assuming $H = 0{\cdot}5$ for oil quenching) were found to be:

1½%Ni–Cr–Mo 0·5 in. 1½%Ni–Cr–Mo + B 4·3 in.

That is, in the 3-in. and 4-in. dia. specimens, the boron-treated steel would be fully hardened, whereas the boron-free steel would be incompletely hardened; the boron-free steel again gave better impact results.

It has previously been stated[9] that for wrought steels the application of boron appeared to be for intermediate section sizes, where boron steels harden right through and boron-free steels do not. The above experiments do not confirm this statement, in so far as cast steels are concerned. However, the results of the above tests should be accepted with caution, in view of some of their conflicting aspects. For example, that the impact properties of the 1¼%Cr–Mo and the 1½%Ni–Cr–Mo did not decrease with increasing section size, was an unexpected result. Even if full hardening is assumed, the impact values should remain constant and not show an increase. Furthermore, in the four cases where subsidiary standard test pieces were machined from the clover leaf itself (1¼-in. fully hardened section), the impact values were lower than those obtained from the simulated 3-in. dia. section.

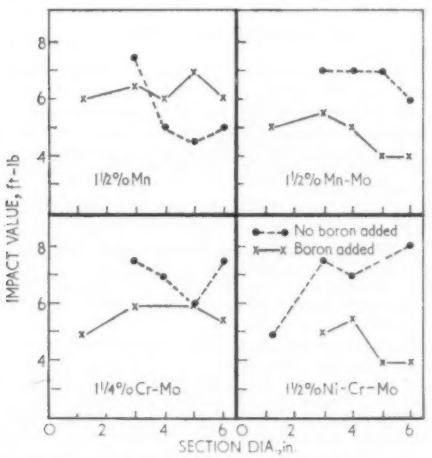


Fig. 8—'Mass effect' impact test results on low-alloy steels with and without boron: subsidiary standard V-notch Charpy tests made at room temperatures

It appears therefore, that either some experimental factor such as oxide formation on the simulation specimens caused scatter of the results, or the Izod machine on which the sub-standard size Charpy specimens were tested was not sufficiently sensitive at low impact energies.

It is concluded that (with the exception of the 1½%Mn steel) the addition of boron tends to lower the impact strength of incompletely hardened cast low-alloy steels. This conclusion is in agreement with that reached on wrought steels[9] but it is felt that confirmation of the results on cast steels could best be made on actual castings of increasing sectional thickness, thus avoiding the use of the oil-quenching simulation technique. The anomalies observed may then be resolved.

Conclusions

The following conclusions are based upon tests carried out on four types of steel (namely 1½%Mn, 1½%Mn–Mo, 1¼%Cr–Mo, 1½%Ni–Cr–Mo), made with and without a boron addition. These were cast from eight separate heats into clover-leaf moulds from a 100-lb basic-lined h.f. furnace, the addition of boron by ferro-boron being preceded by thorough deoxidation with aluminium and ferro-titanium.

(1) The addition of boron increased the hardenability of all the steels tested, as determined by the Jominy end-quench hardenability test, the most outstanding increase being in the case of the 1½%Mn–Mo steel.

(2) The addition of boron did not significantly affect the tensile properties of the fully hardened steels tempered to equal hardness.

(3) The effect of the addition of boron on the impact

characteristics of the fully hardened steels tempered to equal hardness was:

(i) 1½%Mn steel, transition temperature lowered, while maximum impact value remained unaltered
(ii) 1½%Mn–Mo steel, transition temperature lowered, and maximum impact value also lowered
(iii) 1¼%Cr–Mo and 1½%Ni–Cr–Mo steels, transition temperatures raised, and maximum impact values lowered.

(4) In incompletely hardened sections, boron additions tended to lower the impact strength of the 1½%Mn–Mo, 1¼%Cr–Mo, and 1½%Ni–Cr–Mo steels. This effect was not observed on the 1½%Mn steel, and in view of other anomalies observed, all results required confirmation, avoiding the use of the simulation technique.

GENERAL CONCLUSIONS

Section I

Under the experimental conditions described, the low-carbon molybdenum–boron steels gave reproducible and consistent properties. High-tensile strengths were obtained on material in the normalized condition, and L.R.C. tests showed that this improvement may be expected in flat sections of up to 1½ in. thickness. The magnetic properties were fairly good, but did not equal those of carbon steels made to B.S. 1617. The impact properties were not high, but were of the same order as may be expected from carbon steels at a similar tensile level. Since it is likely that a demand exists for steel castings of tensile strength 40–45 tons/in² having a high yield strength and good tensile ductility, together with fairly good magnetic permeability, and good weldability (the latter was not investigated on cast material) it is concluded that for castings not subject to shock loading, low-carbon molybdenum–boron steels can have definite advantages over plain carbon or some types of low-alloy steel castings.

Section II

The addition of boron to low-alloy steels of the 1½%Mn, 1½%Mn–Mo, 1¼%Cr–Mo, and 1½%Ni–Cr–Mo types gave increased hardenability as determined by the Jominy end-quench test, the most marked improvement occurring in the case of the 1½%Mn–Mo steel. In fully hardened sections tempered to equal hardness, the addition of boron did not significantly affect tensile properties, and had an inconsistent effect on the impact properties of the four types of steel. While the impact transition properties of the 1½%Mn and 1½%Mn–Mo steels appeared to be improved by the addition of boron, it is emphasized that the impact transition properties of these steels without the boron addition were inferior to those which may be normally expected. Therefore, the improvement effected by the boron addition may not be significant. In incompletely hardened sections, results of room temperature impact tests were slightly anomalous, and consequently not entirely conclusive, but it appeared that the addition of boron lowered impact strength.

The scale of the investigation of the low-alloy steels, which allowed the examination of only one heat of each steel type, has obviously limited the value of the conclusions. However, it does appear that the improvement in hardenability resulting from boron treatment is not necessarily associated with improved mechanical properties, and the replacement of conventional alloying elements by boron cannot be recommended in the absence of more convincing information.

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APPENDIX I

Steelmaking procedure

In the case of the low-carbon molybdenum–boron steels, the charge of carbon-steel scrap was melted down in 2–2½ h in a 100-lb basic-lined h.f. furnace. After melt-down, about one quarter of the required ferro-silicon was added, followed by pig-iron, ferro-manganese, the remainder of the ferro-silicon, and ferro-molybdenum. Finishing additions were then made in the following order, each addition being made by attachment to a rod which was plunged into the bath and stirred for 30–40 s.

Aluminium	3·2 oz (0·2%)
Ferro-titanium (40% Ti, low carbon)	0·6 oz (0·015%Ti)
Ferro-boron (17% boron)	0·5 oz (0·005%B)
Aluminium	1·6 oz (0·1%)

Slag was skimmed off with a steel rod, and the heats were poured at a temperature of 1580–1600° C into 7½-in. clover-leaf test block moulds, four to a heat.

Procedure was similar in the case of the low-alloy steels, the required ferro-manganese, ferro-silicon, ferro-chromium, ferro-molybdenum, and nickel additions being made after the first addition of ferro-silicon and pig-iron. For the boron-treated heats, finishing additions of aluminium, ferro-titanium, and ferro-boron were made as for the low-carbon molybdenum–boron steels, and pouring was again carried out at 1580–1600° C into 7½-in. clover-leaf test block moulds.

APPENDIX II

Impact testing: method of attaining required test temperature

Charpy impact tests were made over a temperature range of − 155° C to + 135° C. Test pieces were cooled or heated to the required testing temperature as follows:

Below room temperature	Solid carbon dioxide and alcohol mixture
Room temperature to 100° C	Water bath
Above 100° C	Air oven.

Test pieces which were immersed in liquid were allowed 1 h to reach a uniform temperature, while those heated by the air oven were given 2 h.

Austenitic grain-size control of medium-carbon and carburizing steels

By B. L. Biggs

SYNOPSIS

Steelmaking methods used for the production of grain-size controlled medium-carbon and carburizing steels are discussed and the importance of certain critical factors affecting the recovery of acid-soluble aluminium in the steel is considered. Curves are given showing the response of certain steels to grain-size control using aluminium, and the results from two steelmaking processes are compared. Various theories put forward concerning the mechanism of grain-size control are reviewed and details are given of some experimental work carried out on this aspect.

The problem of ingot panel cracking, a defect sometimes found in medium-carbon grain-size controlled steels is also discussed and a possible solution given. 1561

INTRODUCTION

THE ADVANTAGES GAINED from the use of austenitic grain-size controlled steels have for some time been well appreciated by the engineering industries, and consequently appreciable quantities of wrought steel are now supplied to grain-size specifications. These steels have a wide application in the motor industry where their greater uniformity in response to heat treatment and machining operations has been of considerable value, the latter aspect being particularly important where multi-tool machines are operating. The use of fine-grain steels for case-hardening applications has enabled simplified heat treatments to be applied, quenching being carried out direct from carburizing without the need for refining retreatments. Other features include greater freedom from distortion and cracking during heat treatment, shallower hardening characteristics, and greater toughness than coarse-grained steels, and full advantage is taken of these qualities for many applications.

In view of the extensive demand for grain-size controlled steels, the steel industry has developed special steelmaking techniques for their production to ensure that a high standard of quality is maintained. Whilst a considerable number of investigations have been carried out on this subject both in the UK and abroad, most of the published information so far appears to have originated in the USA, and it was

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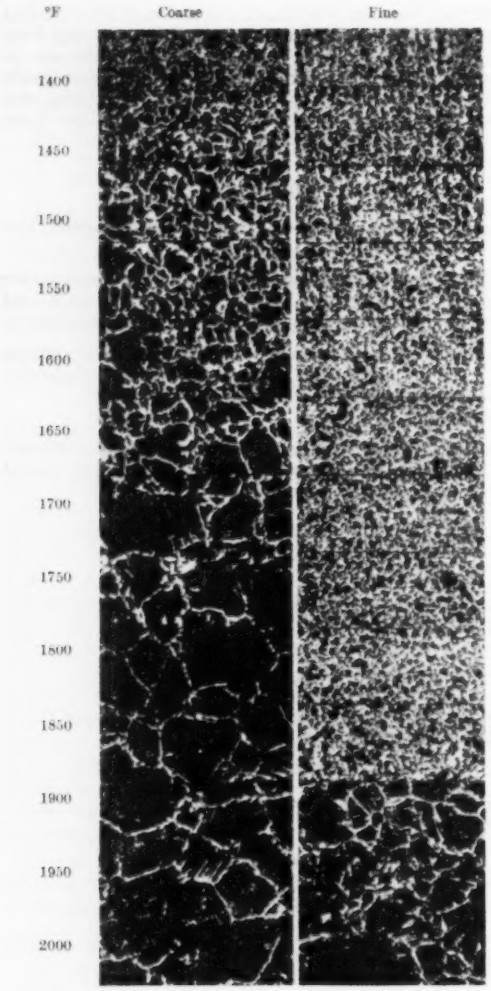


Fig. 1—Composite micrographs showing actual grain-sizes of coarse- and fine-grained steels after normalizing at temperatures shown (Schane[1])

felt that more details of the experience of UK steelmakers in this field of production would be of interest.

This paper deals with the austenitic grain-size produced by heat treatment of wrought steels and is not concerned with the ' as-cast ' grain-size resulting from initial solidification. The techniques employed for the grain-size control of medium-carbon and carburizing steels made by the acid open-hearth, basic open-hearth, and basic electric-arc processes are discussed and comments made on certain critical aspects. The possible mechanism concerned in grain-size control using aluminium and other elements has been the subject of numerous publications. The subject is briefly reviewed and details are given of some relevant experimental work carried out at the author's works. In addition, the factors affecting the incidence of ingot panel cracking, a defect encountered in medium-carbon steels grain-size controlled by the use of aluminium, are discussed and suggestions made for combating this problem.

GRAIN-GROWTH INHIBITORS

When steel is heated above its transformation temperature the ferrite phase is transformed to austenite which possesses an initially small grain-size. In uncontrolled or coarse-grained steels, grain-growth begins as soon as the austenite appears and persists at a more or less uniform rate with increasing temperature. In controlled or fine-grained steels, grain-growth does not begin until the austenite has been heated to a much higher temperature and a critical temperature is reached at which a rapid increase in grain-size occurs. The action of austenitic grain-growth inhibitors is to raise the austenitic grain-coarsening temperature and so called ' fine-grained ' steels are therefore not immune but are more resistant to grain coarsening. An excellent illustration of this has been given by Schane[1] (*see* Fig. 1).

In steelmaking, aluminium is by far the most widely used element for the control of austenitic grain-size because of its great effectiveness and relatively low cost, and for this purpose it can be used in the commercially pure form or in various combinations with other elements such as silicon, manganese, vanadium, calcium, titanium, and zirconium. With its very high tendency to oxidize and its low specific gravity special precautions are needed to dissolve aluminium in steel and whilst it is generally used in the commercially pure form, advantages have been claimed for its use as a combination alloy from the viewpoint of increased freedom from deleterious non-metallic inclusions and improved grain-size control efficiency. Table I gives details of some of the alloys used for grain-size control purposes.

In discussing the effects of grain-growth inhibitors reference can be made to the work carried out by Halley[2] in which the relative efficiencies of titanium, zirconium, and aluminium are compared. Figure 2 shows a graph derived from Halley's work in which curves are given showing the effects of varying residual amounts of titanium, zirconium, and aluminium on the grain-coarsening temperature of En 5 (0·25–0·35%C) type steels. It will be seen that within the range of contents covered for titanium and zirconium (0–0·17%Ti and 0–0·134%Zr), increasing

TABLE I

Alloys used for grain-size control

Alloy type	Approximate content, %						
	Al	Si	Ti	Mn	C	Zr	V
Al	92·0						
SiAl	20·0	40·0					
FeTi (Low C)	8·0	2·5	38·0	1·0	0·10		
FeTi (High C)	2·0	2·5	18·0		6·7		
SiMnAl	20·0	20·0	1·5	20·0			
Silvaz*	6·0	40·0	10·0		1·0	6·0	10·0
SiZr	1·0	50·0				40·0	
FeV	2·0						50·0

* Boron-free grade

residual amounts of these elements raise the grain-coarsening temperature of the steel. Aluminium, however, bears a different type of relationship, in that the grain-coarsening temperature increases up to a maximum acid-soluble aluminium content of about 0·03%, after which further increases in aluminium result in a gradual reduction in the grain-coarsening temperature.

The curves indicate that of the three elements considered, aluminium is the most effective grain-growth inhibitor with residual element contents of less than about 0·035%, whilst above this level of addition, higher grain-coarsening temperatures are obtainable with the use of titanium. Zirconium is shown to be much less effective as a grain-growth inhibitor than either aluminium or titanium. Experiments carried out by Halley also indicate that aluminium-treated steels are less sensitive to the effect of soaking time at temperature than titanium- and zirconium-treated steels, the former steels apparently showing little effect with soaking times ranging from 2 to 8 h. On the other hand, with the latter steels grain-coarsening occurs at lower temperatures as the soaking time increases.

Halley also gave an illustration of the combined effect of aluminium and titanium on the grain-coarsening temperature and this is shown in Fig. 3 in which curves are given for an aluminium-treated En 5 type steel with and without titanium additions. It will be seen that the grain-coarsening temperature is significantly increased by an addition of only 0·005% residual titanium.

Vanadium also acts as a grain-growth inhibitor when added to steel and whilst its value as an alloying element is partly attributed to this feature, generally it is not primarily used for this reason but to achieve enhanced mechanical properties for certain applications. For this purpose higher additions are employed than are used in the case of aluminium for grain-size control.

An important consideration regarding the suitability of the various grain-growth inhibitors discussed is their tendency to introduce undesirable non-metallic

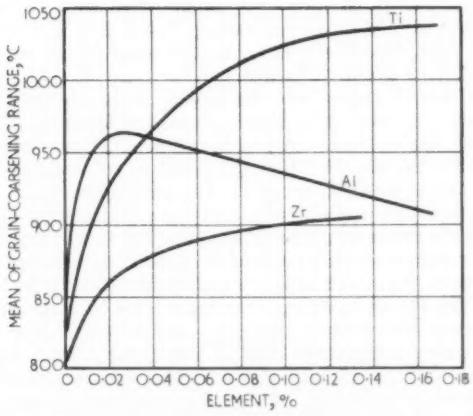


Fig. 2—Effect of titanium, aluminium, and zirconium on grain-coarsening temperature of En 5 type steel (Halley[2])

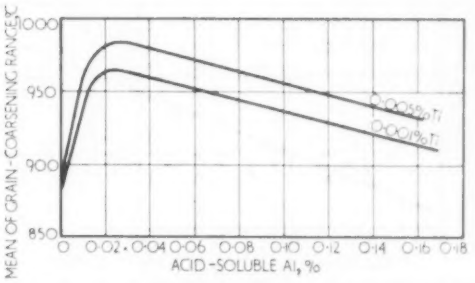


Fig. 3—Effect of aluminium on grain-coarsening temperature of En 5 type steel with varying titanium (Halley[2])

inclusions in the steel, a feature which is of special importance when high-quality steels are being produced. With incorrect steelmaking procedures, such inclusions can be experienced with the use of either zirconium, aluminium, or titanium but of these three elements, aluminium is generally considered to be the least critical in this respect. Titanium tends to produce brittle titanium carbonitride crystals in the steel and therefore its use is not always favoured. For this and other reasons, when titanium is used for grain-size control it is invariably in conjunction with aluminium so that the addition may be kept as low as possible.

Critical aluminium range

In the production of the medium-carbon and carburizing steels it has been found that an approximate range of 0·016–0·035% acid-soluble aluminium is required to ensure fine-grained steel. This is in agreement with the experience of Ressler[3] and other workers. The effective aluminium range varies to some extent according to the grade of steel being produced, since some steels can be satisfactorily controlled with lower residual aluminium contents than required for others, and they also vary in their susceptibility to the reversion to grain-coarsening which is experienced at the higher aluminium-content levels. The author's experience also suggests that basic O.H. steels may be controlled with slightly lower aluminium contents than are required for basic electric-arc steels of similar composition. This aspect is further discussed later.

Rapid assessment of grain-size

In the production of fine-grained steels it is important to know the grain-size characteristics of casts before the ingots are rolled so that any coarse-grained casts which might occasionally be experienced may, if possible, be diverted into other channels of production. The aluminium content of the cast provides a good yardstick for this purpose and for this reason ladle-test samples are analysed on a routine basis for aluminium by the spectrographic method. Although this method is much quicker and cheaper on a routine basis than wet chemical analysis there is a disadvantage in that it reports the metallic, or acid-soluble aluminium, together with the aluminium present as alumina, the latter being ineffective as regards grain-size control. Numerous analyses have, however, been

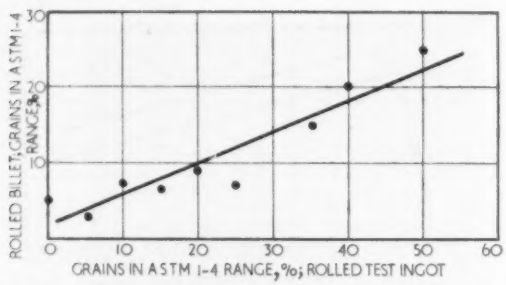
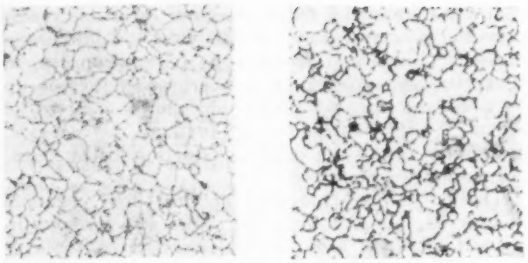


Fig. 4—Relationship between McQuaid–Ehn grain-size results on billets rolled from production ingots, and on rolled 1½-in. square test ingots of En 361 steel



(a) (b)

Fig. 5a—En 355 steel with ASTM grain-size rating of 5% 3–4 and 95% 5–8 ×100
Fig. 5b—En 361 steel with ASTM grain-size rating of 30% 2–4 and 70% 5–8 ×100

carried out for aluminium on aluminium-treated fine-grain steels and this has revealed that the aluminium present as an oxide in these steels remains reasonably constant in the range 0·002–0·005% and these figures can be used as a basis for correcting the spectrographic results to give the acid-soluble aluminium content. It should be mentioned that the spectrographic analysis for aluminium is conducted on transverse sections prepared from the ladle samples, since experience has shown that erratic results can be obtained when the surface of the sample is analysed, owing to the presence of aluminium-oxide-type inclusions. Further, in taking ladle-test samples from aluminium-treated steels losses of aluminium can sometimes occur owing to atmospheric oxidation during the pouring of the samples. Such losses are not usually appreciable, however, but this point should be remembered when using this method of sampling.

Where the aluminium contents of casts are somewhat lower than the desired minimum, the casts concerned are placed in a doubtful category, and in such cases a pilot production ingot is rolled to billet and a McQuaid–Ehn test* carried out for approval before continuing to process the entire cast. An alternative method of obtaining an early assessment of grain-size is the casting of small test ingots which are rolled, carburized, and tested microscopically before processing the production ingots. There can be a difficulty with this method however, in that the McQuaid–Ehn grain-size of such test samples is sometimes found to be somewhat coarser than that of the rolled billet, giving a relationship similar to that shown in Fig. 4. Differences in rolling reduction and rolling temperature and, as mentioned previously, the possibility of lower residual aluminium contents due to losses incurred when sampling, are features which may be important in this connection. In the example shown, the rolling reduction given to the test samples was only 2 : 1 whereas a reduction of about 12 : 1 was applied in the case of the rolled billet. The rolled test-ingot method can be used in conjunction with aluminium analysis to provide an additional check if desired.

Control of various steels

The standard McQuaid–Ehn test is used for the measurement of austenitic grain-size of carburizing and medium-carbon steels at the author's works and the results are classified in accordance with the usual ASTM 1–8 grain-size rating. To facilitate the correlation of grain-size with steelmaking variables and other pertinent factors, attempts have been made to assess the percentage of grains found in specific coarse or fine grain-size ranges and the method has proved valuable for this purpose. Figure 5 gives examples of McQuaid–Ehn tests on two nickel–chromium–molybdenum-type steels which were assessed in this way. Figure 5*a* shows an En 355 steel with an estimated ASTM grain-size rating of 5% 3–4 and 95% 5–8, and Fig. 5*b* shows an En 361 steel with a rating of 30% 2–4 and 70% 5–8.

Table II shows the relationship between spectrographic or total aluminium content and McQuaid–Ehn grain-size obtained on a series of production casts of En 34, En 8M, En 18, and En 361 type steels made by the basic O.H. process. The aluminium analyses were carried out on ladle samples as outlined before and the grain-size determined on rolled billet samples, a procedure which is conducted on all such casts before release. The percentage of grains found to be in the ASTM 1–4 coarse range has been used as a criterion of grain-size control efficiency, and by employing an arbitrary acceptance level of not greater than 15% grains in this category, it will be seen that for effective control a minimum of 0·021% total aluminium is indicated for the En 18 and En 361 steels, whereas minimum contents of 0·016% and 0·011% only are required for the En 8M and En 34 steels respectively.

* The McQuaid–Ehn test[4] which involves carburizing at 925° C for 8 h followed by slow cooling, is employed at the author's works for the assessment of austenitic grain-size. This test was developed in 1922 as a result of an investigation into the effect of steelmaking practice on the hardenability of case-hardening steels. Some engineers also regard a coarse McQuaid–Ehn grain-size as a criterion of easy machinability. Although primarily intended for carburized steels, its use has since been extended to other ranges including deep-hardening grades. It is of special value for steels intended for service in the case-hardened condition but for other steels its interpretation requires some caution since the degree of grain-growth can be influenced by the carburizing conditions applied as well as the temperature and duration of heat treatment. Experience has shown, however, that it provides a satisfactory means of distinguishing between controlled and uncontrolled steels required for many applications and it is therefore widely used as a standard acceptance test.

TABLE II
Total aluminium content of basic open-hearth steels related to McQuaid-Ehn grain-size

Steel type	Total aluminium range, %	No. of casts considered	Percentage of casts in grain-size ranges shown below (Ranges indicate % of grains in ASTM 1–4 category)			
			0–10%	11–15%	16–20%	>20%
En 34 Ni–Mo carburizing	<0·010	13	77·0	7·6	...	15·4
	0·011–0·015	10	100·0	...	...	...
	0·016–0·020	27	100·0	...	...	...
	0·021–0·040	132	98·5	1·5	...	...
	0·041–0·053	12	100·0			
En 8M medium-carbon free cutting	<0·010	12	75·0	8·3	...	16·7
	0·011–0·015	48	87·5	10·4	2·1	...
	0·016–0·020	44	91·0	9·0	...	...
	0·021–0·040	27	85·0	15·0	...	...
	0·063	1	...	100·0	...	...
En 18 medium carbon 1·0%Cr	<0·010	52	61·6	17·3	9·6	11·5
	0·011–0·015	128	73·4	18·0	3·9	4·7
	0·016–0·020	81	77·8	14·8	7·4	...
	0·021–0·035	39	92·5	7·5	...	...
En 361 Cr–Ni–Mo carburizing	<0·010	15	20·0	6·7	6·7	66·6
	0·011–0·015	20	70·0	...	10·0	20·0
	0·016–0·020	39	64·1	25·6	2·6	7·7
	0·021–0·040	113	91·0	9·0	...	...
	0·041–0·058	19	73·6	26·4	...	...

The data from which Table II was compiled is presented in graphical form in Fig. 6. This method of presentation illustrates perhaps more clearly the characteristic response of the various steels to aluminium treatment for grain-size control, each point on the curves representing the mean of several results. The relationships shown by these graphs can be summarized as follows.

En 34 *Ni–Mo steel, carburizing*—Of the four steels considered this is by far the most easily controlled, since it requires the lowest residual aluminium content and also attains the smallest mean grain-size. The curve given indicates a very slight tendency towards a reversion to grain coarsening when the total aluminium content exceeds about 0·020%.

En 8*M steel, medium-carbon free-cutting*—This steel controls almost as easily as the En 34 type but does not appear to attain as fine a mean grain-size.

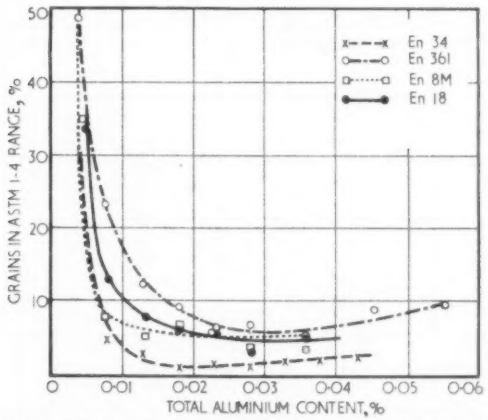


Fig. 6—Effect of aluminium on the McQuaid–Ehn grain-size of En 34, En 8M, En 18, and En 361 steels made by basic O.H. process

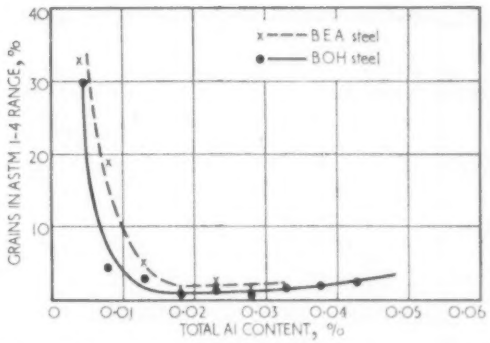


Fig. 7—Effect of aluminium on the McQuaid–Ehn grain size of En 34 steel made by the basic O.H. and basic electric-arc processes

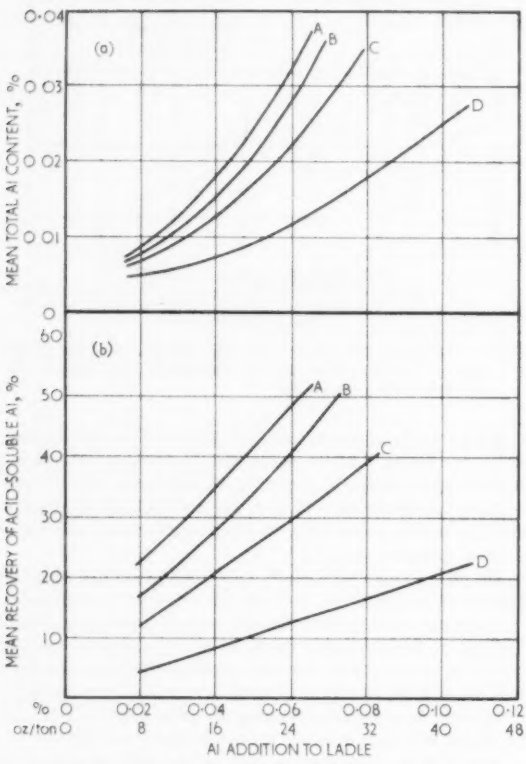
En 18 *steel, medium-carbon* 1%*Cr*—This steel is not as amenable to control as the En 34 and En 8M grades since higher residual aluminium contents tend to be required. However, it can be controlled to give a similar mean grain-size to that obtainable with the En 8M grade.

En 361 *Cr–Ni–Mo steel, carburizing*—This steel is the most difficult to control of the four considered, the highest residual aluminium content being required. However, with suitable residual aluminium contents it controls to a grain-size of fineness similar to that achieved with the En 8M and En 18 grades. A slight reversion to grain-coarsening is indicated when the total aluminium content exceeds about 0·035%.

Production experience has suggested that basic O.H. steels may be somewhat more easily controlled than the equivalent steels made by the basic electric-arc process. In Fig. 7 a comparison is given of the relationship found between McQuaid–Ehn grain-size results and the total aluminium contents of En 34 type steel made by both these steelmaking processes. It will be seen that in the low residual aluminium content range i.e. below about 0·015% the basic O.H. steel tends to attain a slightly finer mean grain-size than the basic electric steel, thus indicating a quicker response to treatment with aluminium. With higher aluminium contents the basic electric steel attains a similar grain-size level as that shown by the O.H. steel. Evidence of a similar relationship has also been found with other steels including the En 18 and En 361 grades and whilst perhaps requiring confirmation it does suggest that a process effect may be operating.

Effect of steelmaking variables

In the case of O.H. steels, aluminium additions for grain-size control are usually made during the tapping of the liquid steel into the ladle, or during casting either directly into the ingot moulds or via the trumpets in the case of uphill pouring. Although lower yields of aluminium are obtained, the former method is usually preferred since it is believed that better uniformity and greater freedom from aluminous-type inclusions are thus achieved. However, in the acid O.H. process it is sometimes preferred to make the aluminium additions to the ingot moulds so as to avoid heavy losses due to the reaction of



Carbon contents, %: A, 0·30–0·40; B, 0·20–0·29; C, 0·13–0·19; D, 0·13–0·40. Special precautions taken against slag contact in A–C only

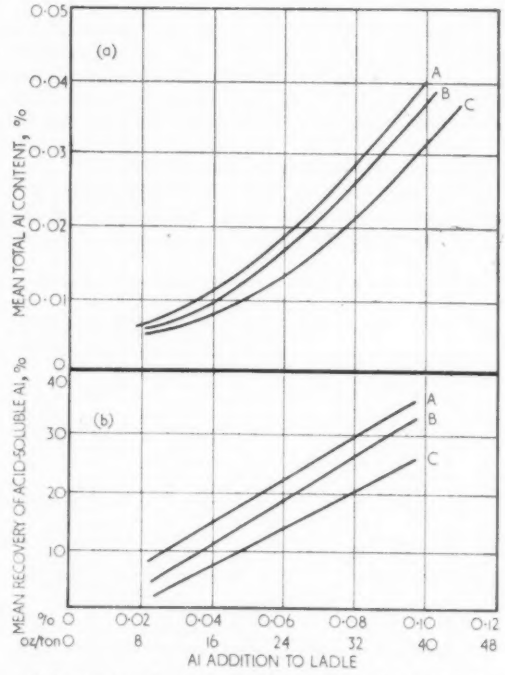
Fig. 8—Effect of amount of aluminium added and carbon-content of steel on the recovery of aluminium in basic electric-arc steels

aluminium with the highly siliceous slag. With basic electric-arc steels the additions are added to the furnace bath, ladle, or ingot moulds. Excessive losses owing to reaction with the slag and possible atmospheric oxidation during tapping can be experienced with additions to the bath and better recoveries are obtained with ladle or ingot-mould additions. Again however, although the highest recoveries are obtainable with mould additions, the addition of aluminium to the ladle is usually preferred to obtain greater homogeneity and freedom from non-metallic inclusions in the steel.

The approximate order of aluminium recovery achieved with basic electric-arc medium-carbon steels using the following methods of addition is: (*a*) bath additions immediately before tapping 15%, (*b*) ladle additions 30–40%, (*c*) mould additions 50–75%.

One of the most important factors in the production of grain-size controlled steels is the control of the state of deoxidation of the steel. It has already been shown that for efficient grain-size control the aluminium content of the steel must be maintained within certain specified limits and it has been found that the recovery of this element can be greatly influenced by the carbon, silicon, and manganese levels, which are important factors governing the oxygen content of the liquid steel. Within the usual limits of composition for medium-carbon and carburizing steels, carbon and silicon have the greater influence in this respect, the effect of manganese being less important. A further critical factor in relation to the control of aluminium content is the way in which the aluminium is added to the steel.

From the data made available by the analysis of ladle-test samples obtained from production casts, curves have been drawn showing the amount of aluminium recovered in the steel with varying additions of aluminium and at different carbon levels. These are given for basic electric-arc and basic O.H. steels in Figs. 8 and 9 respectively. Figures 8*a* and 9*a* show the total aluminium content as determined spectrographically (i.e. including aluminium present as Al_2O_3), while Figs. 8*b* and 9*b* show that the percentage recovery of acid-soluble aluminium in the finished steel increases with the amount of aluminium added and also as the carbon content of the steel increases. Since larger proportions of aluminium will be lost with small additions, and the higher residual oxygen contents of low-carbon steels will also encourage greater losses, this relationship would be expected. In the case of the basic electric-arc steels, it will be seen that a major effect is produced by taking precautions against slag contact to avoid loss of aluminium owing to reaction with the tapping slag. When these precautions are taken the recovery of aluminium is about trebled and is also appreciably more consistent, which of course enables lower aluminium additions to be made.



Carbon contents, %: A, 0·30–0·43; B, 0·20–0·29; C, 0·10–0·19

Fig. 9—Effect of amount of aluminium added and carbon content of steel on the recovery of aluminium in basic O.H. steels

Several methods are available for avoiding slag contact in the electric process, the two perhaps most commonly used being (*a*) plunging the aluminium into the steel attached to the end of suitably designed rods, or (*b*) restricting the size of the furnace tap hole so that the slag layer rises quickly above it during tapping and little slag penetrates into the ladle when the aluminium addition is being made. The latter method has the disadvantages that it encourages greater atmospheric oxidation of the steel, tends to reduce any sulphur removal which occurs during tapping, and also retains slag in the furnace, which can cause fettling difficulties.

A comparison of the results obtained from the basic O.H. and basic electric-arc processes shows that when special precautions are taken against slag contact, in the latter process the aluminium recovery obtained from ladle additions is less variable and about twice that obtained with the former. This is perhaps not unexpected since in the electric process the steel is deoxidized in the furnace whereas in the basic O.H. process the steel is tapped in an unkilled condition and deoxidized in the ladle, which encourages greater oxidation losses of aluminium. With fixed O.H. furnaces the tapping stream is usually kept reasonably free from slag until the final stages, but with tilting-type furnaces unless special precautions are taken some slag may be poured into the ladle during the initial stages of tapping, which can cause high and variable losses of aluminium.

The effect of the carbon content of the finished steel on the recovery of aluminium, is further illustrated in Fig. 10 in which curves are given showing the aluminium recovery obtained with basic O.H. and basic electric-arc steels at varying carbon contents. These curves indicate that the aluminium recovery increases with carbon content at a diminishing rate tending to stabilize at the higher carbon contents. The carbon–oxygen equilibrium curve is shown in Fig. 11 and whilst it is appreciated that equilibrium conditions are rarely experienced in routine steelmaking it will be seen that the recovery curves show a similar characteristic hyperbolic relationship which indicates their apparent association with the residual oxygen-content of the steel.

Steelmaking experience has indicated that the stopping carbon, i.e. the minimum carbon content

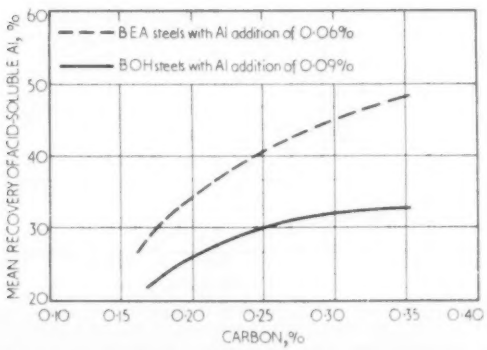


Fig. 10—Effect of carbon content on aluminium recovery in basic O.H. and basic electric-arc steels

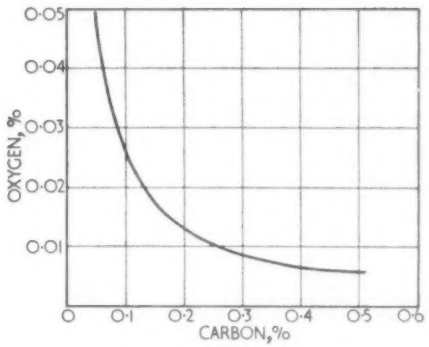


Fig. 11—Carbon–oxygen equilibrium in liquid iron at 1600° C (Marshall and Chipman[5])

attained during the boil, can also affect the aluminium recovery, primarily in the case of the low-carbon carburizing steels. At the low carbon levels the position becomes critical in that small decreases in carbon content can result in relatively large increases in the oxygen content of the steel and it is therefore important in the interests of aluminium recovery, and incidentally freedom from undesirable non-metallic inclusions, that unduly low stopping carbons should be avoided.

The silicon content of the steel also has an important effect on aluminium recovery. Consideration of the silicon–oxygen equilibrium curve given in Fig. 12 shows that as the silicon content of the steel falls below about 0·2% the oxygen content increases fairly rapidly, and in the absence of other strong deoxidizers this can greatly influence the aluminium recovery in the steel and therefore the efficiency of grain-size control. This point is clear in Fig. 13 which shows the efficiency of grain-size control obtained with a basic electric-arc En 34 steel and a basic O.H. En 8M type steel at varying silicon contents, the aluminium grain-size additions being respectively 16 oz/ton and 20 oz/ton to the ladle. From the histograms given it will be seen that in both cases the percentage of coarse-grained casts increases rapidly with silicon contents below about 0·2%, owing to higher oxidation losses of aluminium. For this reason when fine-grained steels are made to specifications which call for limiting silicon contents, careful consideration has to be given to this point when deciding the required aluminium additions, and of course in all cases rigid control over deoxidation procedures must be maintained.

Figure 14 shows the effect of carbon and silicon contents on the grain-size control of a basic electric-arc En 34 type steel. The combined effect of both these elements on grain-size, which reflects the efficiency of the recovery of aluminium in the steel, is clearly seen. The aluminium addition which was used in this instance was obviously inadequate to accommodate the increased oxidation losses experienced at the lower silicon and carbon levels and in addition retain sufficient acid-soluble aluminium in the steel for effective grain-size control.

Table III gives some typical aluminium additions used in the production of basic O.H. and basic

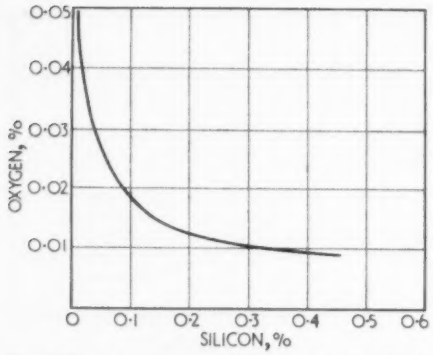


Fig. 12—Silicon–oxygen equilibrium in liquid iron at 1600° C (Gokeen and Chipman[6])

electric-arc grain-size controlled steels required to meet ASTM 5–8 specification.

A further possible cause of variation in the aluminium content of the finished steel is reaction with the slag cover in the ladle during the holding period before casting. Analyses carried out on early and late ladle-test samples taken during casting have shown losses in aluminium up to about 0·003% for 10–15-ton basic electric-arc casts and up to 0·006% for 40–60-ton basic O.H. casts where the holding period in the ladle is much longer. Such losses are, however, by no means consistent but do confirm the desirability of taking late ladle tests for aluminium analysis. When calculating the aluminium additions for grain-size control purposes allowance should be made for possible losses at this stage so as to avoid the risk of obtaining somewhat coarser-grained material from ingots in the latter part of a cast.

Mechanism of grain-size control

Although various theories have been postulated regarding the mechanism by which austenitic grain-size control is achieved with aluminium and certain other elements, a truly satisfactory explanation has yet to be given. The following gives a very brief summary of some of the theories put forward. A fairly widely accepted explanation has been the alumina nuclei theory which attributes the grain-growth inhibiting action of aluminium to the presence in the steel of very finely dispersed particles of alumina, an important factor being said to be the prior state of deoxidation of the steel which affects the particle size of the alumina inclusions. Steels high in oxygen were thought to produce relatively large alumina inclusions which by virtue of their size would tend to rise in the liquid steel instead of forming the desired small nuclei. Conversely, heavily reduced steels were believed to contain too little oxygen for the nuclei to be formed. A further theory proposed by McQuaid[7] is that the action of aluminium is to lower the solubility of carbon in austenite thereby favouring the formation of excess carbides around the austenite grain-boundaries which would tend to retard grain-growth until the carbides were taken into solution at higher temperatures.

From their work on the effect of small additions of vanadium to eutectoid steel it was concluded by Zimmerman, Aborn, and Bain[8] that the grain-growth

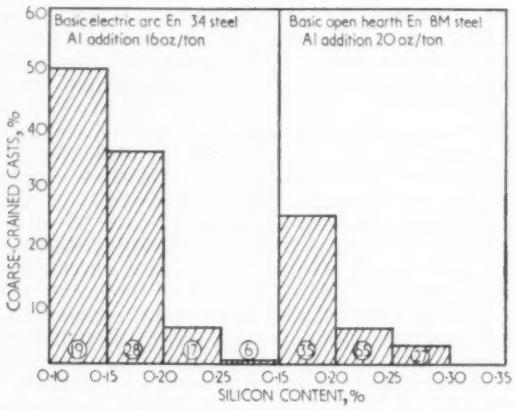


Fig. 13—Histograms showing effect of silicon on grain-size control of basic electric-arc En 34 steel and basic O.H. En 8M steel. Ringed figures denote number of casts examined

inhibiting action of vanadium was due to the presence of relatively stable vanadium-rich carbides and that grain-coarsening only occurred when the carbides were substantially dissolved. Table IV gives details of experimental results obtained by these authors with regard to the grain-size and degree of carbide solution achieved with a straight 0·9% carbon steel and a similar steel plus 0·27%V, when quenched from various temperatures. Other workers, however, have not subscribed to this carbide theory and believe that the formation of vanadium nitrides or oxides may in some way be responsible.

In their notable work, Swinden and Bolsover[9] threw doubt on the nuclei theory and also on the views proposed by McQuaid and considered that the explanation probably lay in the degree of deoxidation of the steel. They compared the alumina contents of fine-

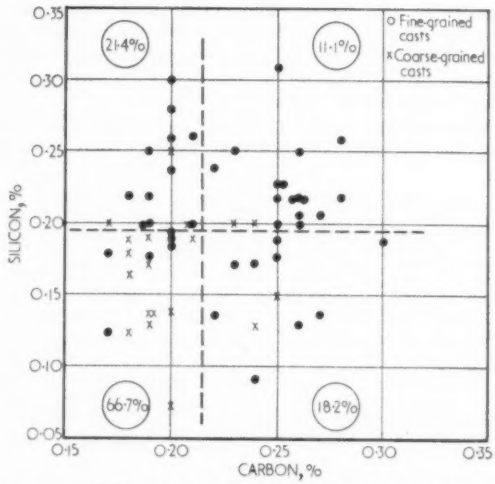


Fig. 14—Effect of carbon and silicon on grain-size control of En 34 type steel made by basic electric-arc process with aluminium addition of 16 oz/ton. Ringed figures give percentage of coarse-grained casts experienced in groups shown

TABLE III

Typical aluminium additions used for grain-size control

Carbon content, %	Aluminium addition to ladle, oz/ton	
	Basic O.H.	Basic electric-arc
0·13–0·20	30–34	20–24
0·21–0·30	26–30	18–22
0·31–0·40	20–24 (17)	14–18
0·41–0·50	20–24 (17)	14–18

Notes: Additions given are for steels with silicon contents in the range 0·20–0·35%
Figures in brackets give additions when an ASTM 3–5 specification is required

and coarse-grained steels and discovered no essential difference but found acid-soluble aluminium contents of 0·02–0·03% in fine-grained steels compared with 0·001–0·007% in coarse-grained steels. It was thought that excess aluminium in itself was not necessarily the controlling factor but that the higher aluminium contents were indicative of a greater degree of deoxidation of the steel. However, an explanation of the mechanism of grain-size control was not offered.

Experimental work carried out by Baeyertz[10] and Sims[11] suggested that aluminium sulphide may be a contributory factor in controlling austenite grain-size. Their work indicated that even in the absence of aluminium, sulphur has a mild grain-refining effect at lower reheating temperatures and that the addition of sulphur to aluminium-treated steels generally gave a higher grain-coarsening temperature than could be obtained with aluminium alone. Figure 15, based on work by Sims,[11] shows the relationship obtained between aluminium additions and grain-coarsening temperature for high-frequency melted 0·4% carbon steels in the wrought condition at three sulphur-content levels, namely 0·012%, 0·024%, and 0·038%.

TABLE IV

ASTM grain-size and degree of carbide solution in 0·9% carbon steel with and without the presence of vanadium*

Quenching temperature, °C	ASTM grain-size		State of carbides	
	Carbon steel	Carbon steel + 0·27%V	Carbon steel	Carbon steel + 0·27%V
760	6	10–11	Partially dissolved	Undissolved
815	6	9–10	Essentially dissolved	,,
870	5	8	,,	,,
925	4	7	,,	Partially dissolved
980	3–4	6–7	,,	,,
1040	3–4	3–4	,,	Largely dissolved
1095	2–3	2–3	,,	Essentially dissolved
1150	2–3	0	,,	,,

* Ref.: Zimmerman, Aborn, and Bain[8]

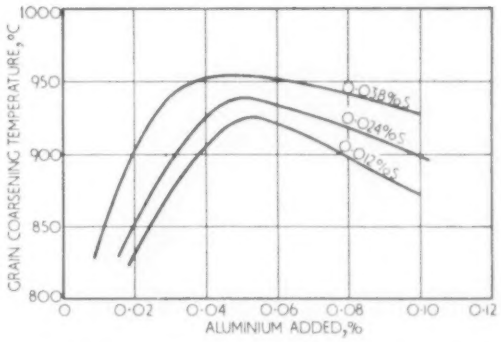


Fig. 15—Effect of sulphur and aluminium on the grain-coarsening temperature of 0·4%C steel (Sims[11])

In more recent years a predominating theory has been that aluminium nitride is the responsible factor in controlling grain-size by the addition of aluminium. Evidence on this aspect was provided by Houdremont and Schrader[12] and other workers and the subject has been more recently dealt with in excellent papers by Bardgett and Gemmill[13] and by Chatterjea and Nijhawan.[14] It has been stated that supporting evidence appears to lie in the fact that aluminium-treated Bessemer steels which are inherently high in nitrogen content tend to possess higher grain-coarsening temperatures than O.H. steels of lower nitrogen content. Furthermore, in addition to aluminium other effective grain-growth inhibiting elements such as titanium, zirconium, and vanadium are also strong nitride formers which suggests that this characteristic may be important. Whilst, however, the precipitation characteristics of aluminium nitride strongly suggest that it could be a pertinent factor in the mechanism responsible for grain refining in aluminium-treated steels, possibly acting as a grain-boundary filament restricting the growth of the austenite grains, there is a school of thought which considers that the amount of nitrogen retained in solid solution may be the critical factor and that a fine grain-size would automatically be experienced in a completely nitrogen-free steel.

In the discussion of Swinden and Bolsover's paper,[9] Herty described the results from two experimental high-frequency furnace casts of medium-carbon steel, one of which was melted and cast in a hydrogen atmosphere and the other in air, and an aluminium addition of 0·10% made in each case, giving residual aluminium contents of 0·082% and 0·071% respectively. After normalizing at 845° C the first cast was found to be coarse-grained and the second fine-grained. Although it was considered then that the fine-grain characteristics of the cast melted in air were probably due to formation of alumina it could possibly be that this cast possessed a higher nitrogen content which would favour the formation of aluminium nitride.

Some experiments have been carried out at the author's works concerning the possible influence of aluminium nitride as a grain-growth inhibitor in aluminium-treated steels, and for this purpose McQuaid–Ehn tests and the necessary chemical and spectrographic analyses were carried out on rolled-

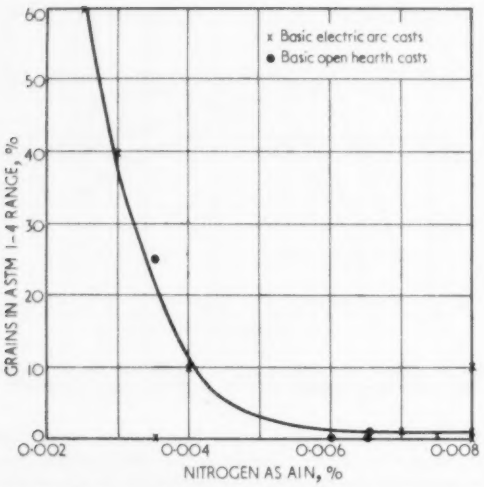


Fig. 16—Relationship found between McQuaid–Ehn grain-size of En 18 steel and nitrogen in the steel present as aluminium nitride

billet samples obtained from a series of basic O.H. and basic electric-arc production casts of En 18 (0·4%C, 1%Cr) steel. The method devised by Beeghly[15] was used for the determination of aluminium nitride and the specimens were analysed in the 930° C (8 h) carburized condition, i.e. the standard treatment for the McQuaid–Ehn test. It was found that a small increase in nitrogen content had occurred during the carburizing treatment but it was not thought that this would necessarily invalidate the results. Figure 16 shows the close relationship obtained between grain-size and the nitrogen in the steel present as aluminium nitride. In later experiments this correlation was repeated using non-carburized specimens in the 930° C (2 h) furnace-cooled condition for the aluminium nitride determinations, which did not reveal evidence of nitrogen increase due to heat treatment and, although showing rather more scatter, a similar relationship to that shown in Fig. 16 was obtained. Only a poor relationship (not shown) was found between grain-size and the total nitrogen content of the steel less the nitrogen present as aluminium nitride, the latter factor being used to give some measure of the nitrogen in solution.

The aluminium nitride in the steel is governed by the aluminium and nitrogen contents and the thermal treatment applied as has been illustrated by the work of Leslie *et al.*[16] Figure 17 shows the close relationship found between the nitrogen present as aluminium nitride in the steel and the aluminium content of the above series of basic O.H. and basic electric-arc casts with similar thermal histories. The curves follow the pattern expected in view of the difference in the inherent nitrogen levels between the two steelmaking processes and show that, with equivalent aluminium contents, much higher aluminium-nitride contents can be obtained with basic electric-arc steels. In general these steels are higher and more erratic in nitrogen content than basic O.H. steels, typical ranges being 0·005–0·007% with basic O.H. and 0·007–0·012% with basic electric-arc, and assuming that the austenite grain-coarsening temperature is in fact affected by the amount of aluminium nitride present in the steel it is likely that this inherent difference in nitrogen content would result in a better response of basic electric-arc steels to grain-size control. As has been mentioned previously however, the correlation of aluminium contents with McQuaid–Ehn tests on production casts has indicated that with residual aluminium contents below about 0·015%, basic O.H. steels tend to show a slightly greater response to grain-size control than basic electric-arc steels even though ultimately with higher aluminium contents equivalent grain-size ratings are obtained (*see* Fig. 7). Interpretation of this is difficult on the basis of the aluminium nitride theory alone and suggests the possibility that other factors inherently associated with these processes such as the difference in sulphur-content levels may be operating. The theory that grain-size control is achieved by the removal of nitrogen from solid solution as a result of the formation of aluminium nitride would perhaps appear to offer a better explanation in this instance. However, if this were so it is difficult to reconcile this with the facts that Bessemer steels of high nitrogen content are reputed to possess relatively high grain-coarsening temperatures, and acid O.H. steels, which are inherently lower in nitrogen content (typical range 0·004–0·005%) than either basic O.H. or basic electric-arc steel, do not appear to be the most readily controlled.

On existing evidence it is concluded that whilst the aluminium-nitride theory may offer a possible explanation for the austenitic grain-growth inhibiting action of aluminium, considerably more evidence will be needed, with particular reference to the action of other grain-refiners such as titanium, zirconium, and vanadium, and the response of steels made by the various steelmaking processes with their characteristic differences before the mechanism responsible for grain-size control can be fully understood. The rapidly expanding volume of work which is being

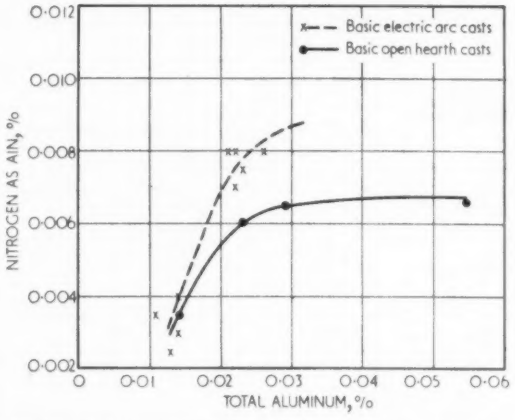


Fig. 17—Relationship between aluminium content and nitrogen present as aluminium nitride in basic electric-arc and basic O.H. casts of En 18 steel (specimens in 930° C (8 h) carburized condition)

carried out in the metallurgical field on vacuum-melted and cast steel may prove very informative in this respect.

INGOT PANEL CRACKING

In the production of grain-size controlled steels a very undesirable ingot defect known as 'panel cracking' or 'centre face cracking' is sometimes encountered. Since most steelmakers engaged in the production of such steels will no doubt be familiar with this defect, details of some experiences with this problem should be of interest.

Panel cracking which is essentially confined to medium-carbon steels takes the form of a longitudinal crack running down the middle of the face or flute of an ingot. It appears to be mainly experienced in the intermediate size range of forging or rolling ingots and is apparently rarely encountered in very large forging ingots. It must not be confused with the interdendritic or 'hot tear' type of longitudinal cracking which is usually found to be responsive to changes in mould design and casting conditions. Panel cracks are not of this nature and evidence suggests that they occur at a much lower temperature, and efforts to eliminate them by alterations in casting technique do not generally prove very effective, although it is claimed that reductions in casting temperature, speed, and careful spacing of moulds in the casting pit to ensure uniform cooling are beneficial. An example of panel cracking in a 14-in. square 24 cwt basic electric-arc En 18 steel ingot is shown in Fig. 18. The defect usually occurs spasmodically and since it almost invariably necessitates the ingot being scrapped it can, if the incidence is large, create serious difficulties with production schedules.

Experience has shown that steels made by the basic electric-arc process are the most prone to cracking, basic O.H. steels being a good deal less susceptible. It is rarely, if ever, encountered in ingots produced by the acid O.H. process. Plain carbon steels in the range 0·4–0·7% C appear to be affected, but there is evidence that the lower limit of carbon is about 0·3% for certain susceptible low-alloy steel grades. Steels particularly prone are the medium carbon grades containing in addition 1%Cr, 1%Ni, or 1½%Mn. The inclusion of 0·2–0·3%Mo in the 1%Cr or 1½%Mn grades eliminates the susceptibility of these steels but it has been found that small residual amounts up to about 0·1% do not have any marked effect.

It appears fairly certain that if the production procedure applied is such that the ingots are stripped and transformed hot direct to the forge or mill-reheating furnaces little cracking is encountered, and that the trouble is mainly experienced where the ingots are allowed to cool after stripping for surface rectification or other reasons. If the latter procedure has to be applied then it is important that other means to eliminate cracking are found.

The defects are usually discovered in the melting shop after the ingots have cooled or at a later stage during machining or deseaming operations. Alternatively they may become apparent during rolling operations as longitudinal cracks in the rolled billets. One or two ingots can be involved, or in very severe cases a complete cast and with square-sectioned ingots cracks can be experienced in one, two, or very occasionally three faces of the ingot. Figure 19 illustrates an etched transverse section taken from a 14-in. square medium-carbon 1%Cr steel ingot showing cracks in two adjacent panels. Here, one of the cracks was not visible on the surface of the ingot, a feature that has frequently been observed in similarly prepared sections taken from other defective ingots, and which suggests that these defects occur internally and propagate towards the surface.

Fig. 18—Panel crack in 14-in. square 24 cwt En 18 (0·4%C, 1·0%Cr) steel ingot × ⅛

The metallographic examination of sections from cracked ingots has so far yielded indefinite evidence

(B)

(A)

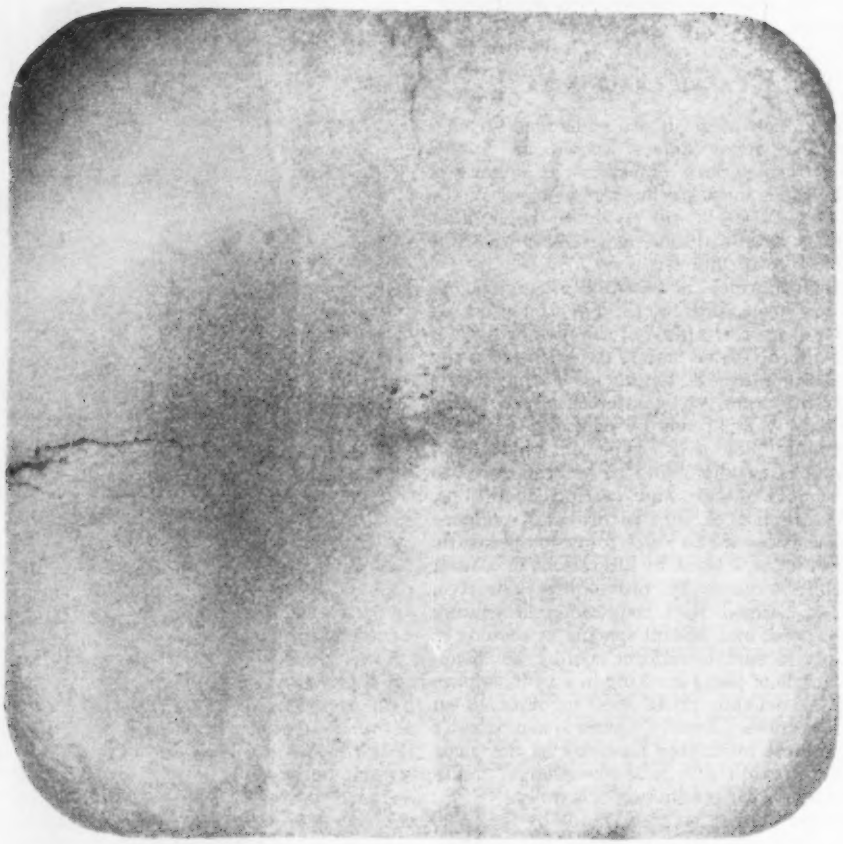


Fig. 19—Transverse section from a position about two-thirds way up an ingot of En 18 (0·4%C, 1·0%Cr) steel, showing panel cracks, (A) visible and the other (B) not visible on the ingot surface × ⅓

of the cause of cracking. Sulphur printing and macro-etching has not revealed any abnormalities associated with the defects. Microscopical examination has revealed that the cracks tend in the main to follow the ferrite network along the prior austenite grain-boundaries, an illustration of this being given in Fig. 20, but also occasionally follow the pearlite-colony boundaries. Only superficial oxidation is usually found on the surface of the cracks and from the manner in which they follow the ferrite network it is considered that they occur below the temperature at which ferrite is precipitated. Further support for this view lies in the fact that cracking occurs in places around the pearlite colonies. Figure 21 is a photograph of an exposed surface of a panel crack showing facets corresponding to primary austenite grains.

On the basis of present evidence it is thought that the cracking occurs as a result of internal stresses set up during the cooling of the ingot and initiates at the prior austenite grain boundaries owing either to the nature of the ferrite network, the presence of a mechanically weak phase, or a combination of these factors. That the defect seems to be essentially confined to steels in the middle carbon range lends some support to the view that a critical distribution or thickness of grain boundary ferrite may be involved, this being controlled by composition, austenitic grain-size, and rate of cooling. This view was supported in a recent paper by Irvine and Pickering[17] in which it was concluded that the cracking was mainly due to the presence of ferrite networks but that impurities such as aluminium nitride might provide the initiating factor for a failure which could propagate through a crack-susceptible structure.

Bjornson and Nathorst[18] reported longitudinal cracks in medium-carbon basic arc steel ingots with abnormally high residual copper and tin contents, namely 0·50–0·68%Cu, 0·016–0·32%Sn. They concluded that the presence of high concentrations of these elements resulted in a weakness in the ferritic network which could lead to cracking under the transformation stresses set up during the cooling of ingots.

Effect of aluminium

An important feature of panel cracking in aluminium-treated steels is its marked relationship with the residual aluminium content of the steel, the incidence becoming greater with increasing aluminium content. The reason for this is not clear but possible explanations are that increasing amounts of aluminium may have an adverse effect on the ferrite hot strength or alternatively encourage the formation of the brittle aluminium-nitride phase at the primary austenite grain boundaries.

The routine analysis of grain-size controlled casts for aluminium has provided extremely useful information on the critical aluminium contents likely to produce panel cracking in a number of susceptible steels made by both basic electric-arc and basic open-hearth processes. Certain authorities still hold the view that panel cracking is associated with 'over killing,' this no doubt being prompted by the fact that treatment with the powerful deoxidizer aluminium produces susceptibility, and further, since basic electric-arc steels which by virtue of the furnace conditions are generally more reduced than basic O.H. steels, are by far the most prone to cracking. Evidence now available that this difference in susceptibility exists at similar residual aluminium levels, and the fact that little trouble is encountered with acid O.H. steels which are inherently well reduced, appears to give little support to this theory.

It has been found that when grain-size controlled medium-carbon steels are made by the basic O.H. process much higher aluminium contents can be tolerated than in the case of basic electric-arc steels, without serious cracking being experienced. Typical figures for a 1%C-type steel are given in Table V.

The actual variable in the aluminium figures quoted in Table V is acid-soluble aluminium since, as was mentioned in the first part of this paper, numerous analyses have shown that the aluminium present in the steel as an oxide remains reasonably constant at 0·002–0·005%. These results indicate quite clearly that factors in steelmaking which influence the recovery of acid-soluble aluminium in the steel by controlling the oxygen content of the bath, e.g. silicon and manganese contents of the steel, type of finishing slag, etc., will have an indirect effect on the incidence of cracking, and also that the effect of other variables such as casting conditions and the thermal treatment of the ingots would be difficult to assess unless correlated at controlled aluminium-content levels. However, the amount of aluminium added does not always provide a correct measure of aluminium present in the steel in view of the considerable variations in recovery which could be experienced owing to varying steelmaking conditions.

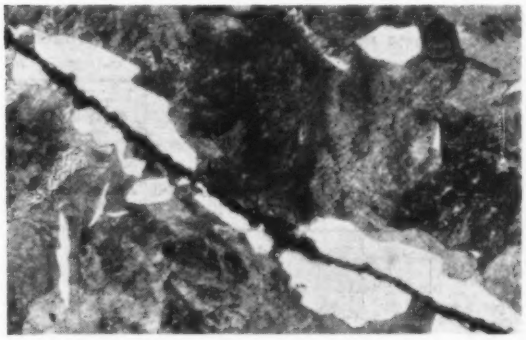


Fig. 20—Crack following ferrite network at prior austenite grain boundary in En 12 (0·4%C, 0·8%Ni) steel ingot × 170

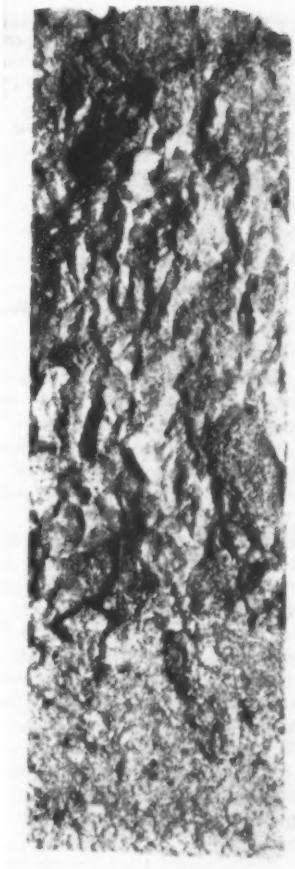


Fig. 21—Inner surface of panel crack showing faceted appearance, actual size

Because of the critical influence of aluminium it is naturally preferred to make grain-size controlled steels, which are very susceptible to panel cracking by the basic or acid O.H. processes, which allow greater tolerance in residual aluminium content. When basic electric-arc steels are required, very great care has to be taken in controlling the aluminium content if serious cracking is to be avoided, and it is very much a compromise between obtaining effective grain-size control and the avoidance of panel cracking.

Attention has been given to ascertaining possible reasons for the difference in susceptibility shown by steels produced by the basic O.H. and basic electric-arc processes, also bearing in mind that little trouble

TABLE V

Incidence of panel cracking in 14 in. square En 18 steel ingots related to aluminium content

Total aluminium content, %	Ingots showing panel cracks, %	
	Basic O.H.	Basic electric-arc
<0·015	0·4	3·0
0·016–0·020	1·7	31·4
0·021–0·025	20·8	44·0
0·026–0·030	35·8	N.D.
0·031–0·045	40·3	71·0

is encountered with acid O.H. steels. There are of course several characteristic differences between steels made by these processes including variation in residual oxygen, hydrogen, sulphur, and phosphorus and in standard of cleanness, but one feature which appeared to be of possible significance in view of the influence shown by aluminium was the inherent difference in nitrogen levels.

It is now fairly well established that the presence of critical amounts of aluminium nitride precipitated at the primary austenite grain boundaries may, with certain carbon and low-alloy steels, produce an intergranular form of fracture often referred to as 'rock candy' fracture. This type of fracture which has been investigated by Lorig and Elsea[19] and several other workers, is associated with poor ductility and low impact properties and it has been reported that in certain cases it may result in the cracking of castings in the mould under the internal stresses set up during cooling. The appearance of this type of fracture is in many respects very similar to that obtained in panel cracking.

It has been postulated[19] that the susceptibility to intergranular fracture depends on the aluminium-nitride content of the steel, this being controlled by the amounts of aluminium and nitrogen present and the rate of cooling, rapid cooling retarding precipitation which occurs principally in the range 980–700° C.

In Table VI the nitrogen and aluminium contents of a series of basic O.H. and basic electric-arc casts of En 18 steel have been correlated with the incidence of panel cracking experienced. This correlation shows the nitrogen level of the O.H. casts to be in the range 0·005–0·006%, and for the electric-arc casts 0·007–0·012% and indicates that increasing aluminium and nitrogen contents tend to increase the incidence of cracking. It should be mentioned here that the nitrogen contents of similar grade steels produced in the acid O.H. where cracking is very rarely experienced lie in the range 0·004–0·005%, i.e. slightly lower than obtained with the basic O.H. process. Analysis by the Beeghly[15] method has confirmed the presence of aluminium nitride in cracked ingots and since the amount present will be related to the aluminium content of the steel it would be expected that casts high in aluminium will also tend to have a relatively high aluminium-nitride content (*see* Fig. 17).

Table VII gives details of hot tensile tests carried out on a series of experimental medium-carbon steel casts with varying aluminium and nitrogen contents. For the purpose of this experiment test bars were cast in sand moulds from a 20-lb h.f. furnace. The specimens were tested in the as-cast condition at 600°, 700°, and 800° C after soaking for 1 h at temperature.

The results indicate that the higher aluminium contents, namely 0·058 and 0·073%, have caused a lowering of the hot ductility of the steel at temperatures of 600° and 700° but not at 800° C. With regard to the effect of nitrogen content there is some indication that when high aluminium contents are accompanied by high nitrogen contents a further reduction in ductility is experienced at the lower temperature.

Experiments carried out in the author's works and elsewhere have indicated that whilst probably the best method of avoiding the occurrence of 'rock candy' type fracture is to ban the use of aluminium altogether, it may possibly be minimized in steels treated with aluminium if a prior addition of titanium or zirconium is made to enable these elements to combine preferentially with nitrogen and so reduce the chances of forming aluminium nitride. Woodfine[20] arrived at the latter conclusion in his work on intergranular fracture in steel castings. From the evidence obtained it was considered that although by no means certain, there was a possibility that aluminium nitride was a contributory factor in panel cracking, and it was therefore decided to carry out production trials with medium-carbon grain-size controlled steels using titanium as a prior nitride former.

Titanium was chosen since it has less affinity for oxygen than aluminium whereas zirconium the probable alternative, although perhaps a stronger nitride former, has an even greater affinity for oxygen and can be subject to severe oxidation losses when added to liquid steel.[21] Further, since titanium also acts as a grain-growth inhibitor this enables the aluminium addition to be reduced, a procedure that is sometimes adopted for purely grain-size control purposes and which it was considered would be a further desirable step to take.

A series of production basic electric and basic open-hearth casts of medium-carbon steel of the highly susceptible 1%Cr and 1½%Mn grades has been made using a prior titanium addition in the form of ferro-

TABLE VI

Incidence of panel cracking in 14 in. square En 18 steel ingots related to aluminium and nitrogen contents

Nitrogen content, %	Total Al: 0·005–0·015%		Total Al: 0·016–0·020%		Total Al: 0·021–0·025%	
	No. of ingots examined	% Cracked	No. of ingots examined	% Cracked	No. of ingots examined	% Cracked
0·005–0·006 (Basic OH steel)	128	Nil	77	6·5	128	18·8
0·007–0·009 (Basic E-A steel)	28	Nil	43	28·0	54	37·0
0·010–0·012 (Basic E-A steel)	14	Nil	82	39·0	29	53·0

TABLE VII

Details of hot tensile-test results on medium-carbon steels with varying aluminium and nitrogen contents

Cast No.	Composition, %					600° C			700° C			800° C		
	C	Si	Mn	Al	N_2	U.T.S.	El.	R.A.	U.T.S.	El.	R.A.	U.T.S.	El.	R.A.
01726B	0·34	0·21	0·89	0·003	0·010	15·0	40	67	9·75	56	36	7·5	61	98
01726E	0·32	0·30	1·22	0·010	0·018	18·0	28*	54	N.D.	N.D.	N.D.	N.D.	N.D.	N.D.
01726D2	0·35	0·29	1·24	0·014	0·007	21·0	34	55	10·0	48	73	7·4	63	100
01726C	0·35	0·37	1·06	0·058	0·015	16·0	11	6	9·0	14*	3	8·25	70	93
01726A2	0·34	0·36	0·98	0·073	0·005	16·0	17*	28	9·6	18*	4	8·25	61	70

* Test piece broken outside middle half N.D. = Not determined

titanium followed by a reduced addition of aluminium, and whilst insufficient casts have been produced to enable a definite conclusion to be drawn the results so far appear very promising in that in every case satisfactory grain-size control has been achieved and the incidence of panel cracking very greatly reduced. With the electric-furnace process the titanium additions can be added to the furnace bath or in the ladle and, as in the case of aluminium, special precautions have to be taken to avoid excessive losses due to reaction with slag. For the open-hearth process the additions are made in the ladle. The titanium addition employed for this method of grain-size control is about 0·04%.

As mentioned earlier, a disadvantage that can arise with titanium-treated steels is the tendency for titanium carbonitride crystals to be formed which can be undesirable when ease of machining or a high standard of cleanness is essential. Careful consideration may have to be given to this aspect when deciding the suitability of this method. The severity of these inclusions is, however, affected by the steelmaking procedure employed and therefore may be minimized to a large extent by the adoption of a suitable steelmaking technique. No difficulties have so far arisen in this direction with any of the casts made by the procedure discussed. Table VIII gives details of some of the casts produced using titanium and aluminium for grain-size control purposes and includes the results obtained concerning grain-size and incidence of panel cracking. The ingots from these casts were stripped hot and air cooled after casting. Later they were taken from stock, preheated and descamed then reheated for rolling. All the casts of which details are given were free of cracking even though two casts possessed dangerously high aluminium contents, i.e. 0·034% and 0·027%.

Table IX gives details of the results obtained with basic O.H. En 18 steel ingots grain-size controlled with titanium and aluminium and, for comparison, the results achieved from similar casts treated solely with aluminium.

Should the procedure outlined prove to be completely effective it is appreciated that this does not necessarily confirm the validity of the aluminium-nitride theory in relation to panel cracking, and that other possible factors concerning the hot strength of ferrite need to be investigated. The fact that suitable

TABLE VIII

Details of grain-size controlled casts using titanium and aluminium

Cast	Process	Composition, %						Ladle addition, %		Steel content, %		McQuaid–Ehn grain-size of rolled billet	Incidence of ingot panel cracking
		C	Si	S	P	Mn	Cr	Ti	Al	Ti	Al		
A (En 15)	Basic E-A	0·35	0·33	0·012	0·021	1·54	...	0·075	0·041	0·040	0·016	Av. 6, range 5–8	Nil
B (En 18)	,,	0·36	0·31	0·013	0·023	0·72	1·03	,,	,,	0·025	0·014	,,	,,
C ,,	,,	0·35	0·23	0·015	0·022	0·68	1·05	,,	,,	0·036	0·013	,,	,,
D ,,	,,	0·38	0·30	0·018	0·019	0·72	1·02	,,	,,	0·030	0·009	,,	,,
E ,,	Basic O-H	0·39	0·31	0·031	0·026	0·72	1·09	0·09	0·06	0·060	0·034	,,	,,
F ,,	,,	0·42	0·27	0·026	0·034	0·79	1·07	0·05	0·04	0·031	0·014	,,	,,
G ,,	,,	0·36	0·28	0·030	0·020	0·75	0·90	,,	,,	0·025	0·010	,,	,,
H ,,	,,	0·36	0·28	0·027	0·029	0·75	0·95	0·04	0·04	0·002	0·027	,,	,,
I ,,	,,	0·35	0·31	0·027	0·026	0·70	0·91	,,	,,	0·018	0·010	,,	,,
J ,,	,,	0·34	0·25	0·034	0·030	0·79	1·00	,,	,,	0·025	0·012	,,	,,
K ,,	,,	0·39	0·31	0·032	0·031	0·80	1·05	,,	,,	0·022	0·016	,,	,,
L ,,	,,	0·39	0·31	0·035	0·032	0·78	1·00	,,	,,	0·023	0·019	,,	,,
M ,,	,,	0·38	0·26	0·038	0·036	0·73	1·02	,,	,,	0·022	0·015	,,	,,

All ingots 14-in. square, bottom poured, stripped hot and air cooled

TABLE IX

Panel cracking in basic open-hearth En 18 steel, 14 in. square ingots related to grain-size control procedure

Total aluminium content, %	Casts controlled with Al		Casts controlled with Ti + Al	
	No. of ingots considered	% Cracked	No. of ingots considered	% Cracked
≤0·015	2517	0·4	1596	Nil
0·016–0·020	1449	1·7	689	,,
0·021–0·025	691	20·8	203	,,
0·026–0·030	231	35·8	82	,,
Totals	4888	5·4	2570	Nil

additions of the strong ferrite strengthener molybdenum provides immunity appears to be significant in this respect. In addition to its strong nitride-forming quality, titanium can also act as a ferrite strengthener[21] and it is possible that some benefit may be gained in this direction. This element certainly does not appear to have the adverse effect on cracking shown by aluminium since no difficulties have been experienced with residual titanium contents of up to 0·06%, this being the maximum content experienced with the casts treated so far. On the other hand it may finally be proved that its value depends on its function as a grain-growth inhibitor, so that lower aluminium contents may be employed which are outside the critical range which causes panel cracking, although this seems doubtful. These are questions which remain unanswered but from all these considerations it would appear that the choice of titanium is probably a suitable one and so may offer a possible means of overcoming the problem.

In view of the promising results achieved with titanium, the use of other nitride formers, i.e. zirconium and vanadium, is being investigated to compare their effectiveness.

SUMMARY AND CONCLUSIONS

Grain-size control

Grain-size control in the production of medium-carbon and carburizing steels is achieved by the addition of suitable elements which have the effect of increasing the austenite grain-coarsening temperature. The most effective element is aluminium, which is added either in the commercially pure form or in combination with other elements such as silicon, manganese, vanadium, calcium, zirconium or titanium. Higher grain-coarsening temperatures are obtainable if titanium is used in conjunction with aluminium, but its tendency to form titanium carbonitride inclusions may be a disadvantage in steels where a high standard of cleanness is essential.

It has been found that in general, acid-soluble aluminium contents within the approximate range 0·016–0·035% are required to produce fine-grained steel to the ASTM 5–8 specification. The effective aluminium range varies to some extent according to the type of steel being made, the chromium-bearing steels requiring somewhat higher aluminium additions than plain carbon steels. Of the steels considered, the En 34 nickel–molybdenum grade was the most easily controlled.

Production experience suggests that basic O.H. steels are somewhat more amenable to grain-size control than basic electric-arc steels of similar composition.

Careful control of the steelmaking practice is essential in the production of fine-grained steels. The state of deoxidation has a marked influence on the amount of aluminium retained in the steel and in this respect the carbon and silicon contents are particularly important. The amount of aluminium recovered increases with the increase in the amount of these two elements and due allowance must be made for this in fixing the aluminium additions. A further important factor is the manner of adding the aluminium and care must be taken to avoid contact with slag. Slight losses of aluminium may sometimes occur during the holding period in the ladle and unless allowance is made for this there may be a tendency for coarser-grained materials to be experienced in the latter part of the cast.

It is important to know the grain-size characteristics of casts before rolling and this may be assessed by either (*a*) the routine analysis for total aluminium content, or (*b*) McQuaid–Ehn tests on material rolled from small pilot ingots.

Experimental work carried out on rolled-billet samples of aluminium-treated En 18 steel indicates a relationship between McQuaid–Ehn grain-size and the aluminium-nitride content of the steel. This is in line with the theory put forward by other workers that the grain-growth inhibiting action of aluminium is due to the formation of aluminium nitride rather than alumina nuclei. Other grain-growth inhibitors such as titanium, vanadium, and zirconium are also nitride formers and it seems likely that their effectiveness is associated with this characteristic. However, it is felt that much more evidence is required before the mechanism of grain-size control can be fully explained. The rapidly expanding volume of work now being carried out in connection with vacuum melting and casting should prove very informative in this respect.

Ingot panel cracking

Ingot panel cracking or centre face cracking as it is sometimes called is a defect which is experienced with medium-carbon steels when treated with aluminium for grain-size control. Since any ingot suffering from this defect almost invariably has to be scrapped it can create a serious production problem if the incidence is large. Evidence suggests that the cracks occur internally and propagate to the surface as a result of internal stresses set up during the cooling of the ingot. Only slight oxidation is usually found on the surface of the cracks and since they tend in the main to follow the ferrite network at the prior austenite grain-boundaries it is concluded that they occur at a temperature below that at which ferrite is formed.

Experience has shown that basic electric-arc steels are more prone to panel cracking than basic O.H.

steels and that the defect is rarely experienced in the case of the acid O.H. process.

A significant relationship has been found between the incidence of cracking and the aluminium content of the steel, the incidence increasing as the aluminium content increases. Higher aluminium contents can be tolerated without experiencing cracking in the case of basic O.H. steels than with basic electric-arc steels of equivalent composition.

The reason for the influence shown by aluminium is not clear but possible explanations are that the presence of increasing amounts of this element in the steel may (*a*) produce an adverse effect on the ferrite hot strength, or (*b*) encourage the formation of a brittle aluminium-nitride phase at the primary austenite grain-boundaries, conditions which could favour cracking under the transformation stresses encountered during cooling of the ingot. Since the more susceptible basic electric-arc steels possess a higher inherent nitrogen content than the basic O.H. or acid O.H. steels the problem was pursued particularly from the aspect of the aluminium-nitride theory.

Production trials were carried out with highly susceptible basic electric-arc and basic O.H. steels of the medium-carbon 1·0%Cr and 1½%Mn grades in which titanium was used in conjunction with aluminium for grain-size control purposes with a view to it acting as a prior nitride former and so minimize the chances of aluminium nitride being formed in the steel. Furthermore, since titanium acts as a grain-growth inhibitor, it should permit the use of lower aluminium additions. This procedure has so far proved very successful in that satisfactory grain-size control has been achieved and the incidence of panel cracking greatly reduced.

ACKNOWLEDGMENTS

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The influence of tempering-time on some of the mechanical and physical properties of steel

By J. L. Aston, M.A., Ph.D.

SYNOPSIS

An apparatus is described for the uniform heat treatment of small steel specimens in periods of a few seconds. It is confirmed that the mechanical properties of hardened and tempered steels can be represented fairly accurately by linear functions of the tempering temperature and the logarithm of the tempering time, but generally with a small interaction between these variables. If two specimens of the same steel are tempered to the same strength by different tempering treatments, they do not necessarily have the same ductility. For the low-alloy steels examined, it is usually found that the specimen which is tempered for the shorter time at the higher temperature possesses the better ductility. This is largely because temper-embrittlement is less. Line-breadth measurements on the (310) X-ray diffraction line from the ferrite indicate no significant difference between specimens of the same hardness, even though there is a difference of 1 : 20000 in the tempering time. 1467

TABLE I

Composition of steels tested, and hardening and quenching treatments

Steel	Composition, %								Hardening Temp., °C	Quenching Medium
	C	Mn	Ni	Cr	Cu	Si	S	P		
A	0·38	0·53	3·43	0·12	...	0·25	0·011	0·008	880	Oil
B	0·40	1·40	0·15	0·06	0·21	0·24	0·046	0·024	840	Water
C	0·47	0·30	0·05	0·05	...	0·08	0·009	0·015	840	Water
D	0·46	0·87	0·18	1·00	...	0·29	0·030	0·028	900	Salt at 200° C

GENERAL INTRODUCTION

ALTHOUGH A GREAT DEAL of attention has been given to the temperature of heat treatment, less has been given to the duration of heat treatment. Treatment time is important not only for furnace efficiency but also in the application of induction heating, where the process is inevitably carried out in a period of seconds or a very few minutes.

Several workers have investigated the effect of time and temperature of tempering on hardness,[1] with the implication that, for each steel, a particular hardness will always have associated with it the same set of values for the other mechanical properties. Other mechanical properties have been measured on steels tempered for various periods[2] but the shortest tempering times were of several minutes duration. It has been suggested[3] that rapid tempering may give a better compromise between strength and ductility than is obtained by furnace tempering.

In the work reported here, an attempt has been made to determine the change of properties during the tempering of steel, with special reference to periods of a few seconds. The apparatus described was later used to investigate the progress of the alpha–gamma change in the hardening reaction.

DESCRIPTION OF APPARATUS

Electric-resistance heating was necessary to give the required heating rate (500–1000°C/s). Since uniform heating was essential, high-frequency current with inductive coupling could not be used because of the 'skin-effect' phenomenon. Instead, 50-cycle alternating current was passed directly through the specimens, which were in the form of bars, about 6 in. long and ¼ in. dia. The necessary current of about 1 500–2 000 A at about 10V was obtained from a tapped transformer. Contact with the ends of the specimens was made through massive copper blocks. The bars were quenched *in situ* with two wide jets of water which played on opposite sides of the specimen, to keep distortion to a minimum.

The heating period and the delay before the beginning of the quench were controlled by electronic timers, working on the condenser discharge principle.

For the tempering experiments, it was sufficient to measure the maximum temperature reached. Chromel and alumel wires of 30 s.w.g. were independently spot welded to the specimens midway between the copper terminal blocks. In experiments with tempering times of 5 s and longer, a Negretti and Zambra quick-reading potentiometer was used to measure the e.m.f. produced. When the tempering time was shorter, a sensitive and quick-acting moving coil meter was used. For the shortest tempering times, a cine camera was used to follow the meter movements.

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EXPERIMENTAL PROCEDURE

Heat treatment

Specimens to be tempered electrically were 6 in. long by ¼ in. dia. and were hardened as shown in Table I, which also shows the analyses of the steels used. Thermocouple wires were attached to a specimen, which was then clamped in the copper blocks, and the timers were set to appropriate heating and delay times. Because of small variables, especially the contact resistance, it was not possible to predetermine accurately the temperature which would be reached, but it was possible to treat a series of specimens to gradually increasing temperatures. The heating time was generally less than 1 s and the delay, or 'soaking' time, from 1 to 5 s. To define accurately these short tempering times it was essential to end the tempering period by quenching rather than by air cooling. Because of this, and the need to maintain uniform conditions, all material was quenched from the tempering temperature irrespective of the tempering time.

Specimens for furnace tempering were 4 in. long, and were tempered in molten salt for temperatures less than 500° C, in lead for short times at higher temperatures, and in a tube-furnace for longer periods at a higher temperature. In the case of the specimens tempered for only a few minutes, it was necessary to make a fairly accurate allowance for the time taken in heating to temperature. The heating time was determined, for the various conditions that were used, by means of a hollow specimen containing a thermocouple.

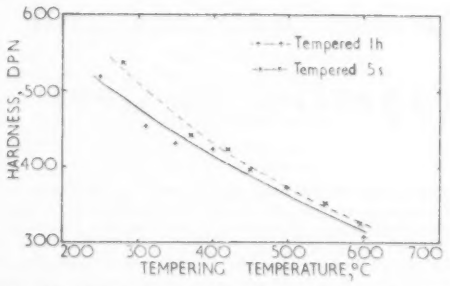

Fig. 1—Hardness *v.* tempering temperature, steel *A* (3·5%Ni, 0·4%C)

Mechanical testing

The ends of the electrically tempered specimens were discarded, as they had been kept cold by the copper blocks. The bars were then turned parallel, and a central length of $\frac{3}{4}$ in. taken down to about 0·18 in. dia. ($\frac{1}{40}$ in² cross-sectional area). The specimens were pulled to destruction, mostly on a 7 000-lb Buckton machine, and the usual tensile-test figures were obtained.

RESULTS OF MAIN SERIES OF TEMPERING EXPERIMENTS

In the first experiments on the 3%Ni steel *A*, one series of test-pieces was tempered for 5 s, and a second series for 1 h. Figure 1 shows the relationship between hardness and tempering temperature for the two tempering periods. These curves are similar in shape, with an approximately uniform displacement along the temperature axis. The U.T.S./tempering-temperature curves are of the same shape.

Figure 2 shows that tempering time has a more marked influence on the ductility than on the hardness. The rise in ductility associated with increasing tempering temperature occurs at a lower temperature in the case of the specimens tempered for the shorter period. No satisfactory explanation of this observation can be offered.

In engineering practice, the heat treatment employed is quite unimportant compared with the properties obtained. The results have therefore been plotted again in Fig. 3, with U.T.S. and reduction in area along the axes. It is evident that a useful improvement in properties may be obtained by tempering for only a short time. It was observed that this improvement in reduction of area value with short tempering time was associated with deformation after the maximum load had been reached.

The 1%Mn steel *B* was tested in the same way, and the strength/ductility relationships for the two tempering times are shown in Fig. 4. The improvement with a short tempering time is smaller than for the nickel steel, but is reproducible.

When these results had been obtained, it was decided to extend them to a greater range of tempering times. It has been shown[1] that the hardness decreases linearly with the logarithm of the tempering time, and this was confirmed. It was also confirmed that strength decreases similarly. However, the ductility did not change in an exactly corresponding way, i.e. it did not increase uniformly with the logarithm of the tempering period. The deviation from the anticipated results was especially marked in the first minute.

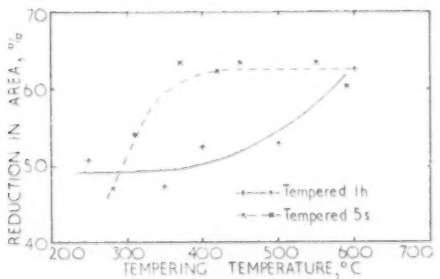

Fig. 2—Ductility *v.* tempering temperature, steel *A*

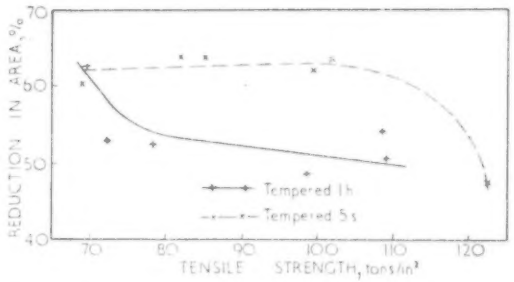

Fig. 3—Strength *v.* ductility after tempering, steel *A*

The results could be presented as a series of families of curves, showing the dependence of strength or ductility on either tempering time or tempering temperature, the other being held constant. Instead, they are presented as 'contour maps' of the hardness–time–temperature and ductility–time–temperature surfaces. Figures 1 and 2 are vertical sections of such surfaces.

Surfaces were fitted mathematically to the results obtained by tempering for various times and at various temperatures. They were fitted by the method of least squares, the mathematical form being selected by inspection of the results, and not by theoretical reasoning, i.e. the surfaces are empirical. Product and square terms were included if their coefficients were statistically significant.

The hardness and ductility of steel *B* after tempering under a wide range of conditions are shown in Fig. 5. Each line in one family of curves joins points of the same hardness on the 'hardness surface'; each line in the other family joins points of the same ductility on the 'ductility surface.' Hardness was plotted in preference to strength because there is a smaller random experimental error in its determination, and the two are closely correlated. It is evident that, for this steel, time and temperature of tempering are not completely interchangeable, i.e. it is not true that if two sets of tempering conditions give the same strength, they will also give the same ductility. If it is required to temper to a definite strength, it will be possible to achieve a greater ductility with that strength if a short time of tempering is used. The improvement is greater toward the lower left-hand part of the diagram, i.e. for the hardest condition, where the two sets of 'contour lines' cross at the greatest angle.

Several specimens were examined microscopically. The structures observed were typical of those which usually occur in a hardened and tempered steel, having regard to the hardness of the various specimens. Micro-examination of the other steels gave the same result, except as noted later.

Two further steels, one a non-alloy medium-carbon steel (*C*) and the other a medium-carbon steel with 1% Cr (*D*) were treated similarly, and the results are shown in Figs. 6 and 7.

Figure 6 shows that for the plain carbon steel, when the hardness is not less than about 350 D.P.N., the

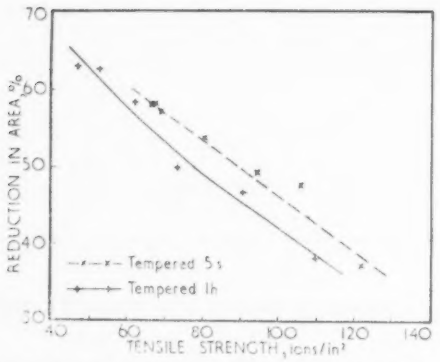

Fig. 4—Strength *v.* ductility after tempering, steel *B* (0·4%C, 1·4%Mn)

two sets of contour lines are substantially parallel, i.e. for any selected hardness the ductility will not depend on the particular combination of tempering time and temperature used to obtain that hardness. However, for softer material, the longer the tempering time (with an appropriately lower temperature) that is used to obtain a required hardness, the higher will be the ductility obtained in conjunction with it.

The results in Fig. 7 for the 1%Cr steel *D* show that except for the softest material (less than about 250 D.P.N.) the more rapid the tempering, the better the compromise between strength and ductility. It will be shown that the advantage is larger with larger specimens, and corresponds to a substantial difference in notched bar value.

X-RAY DIFFRACTION MEASUREMENTS

The results described above show that it is possible for two pieces of the same steel to be tempered to the same hardness, and yet have different ductilities. In an attempt to determine the cause of this difference,

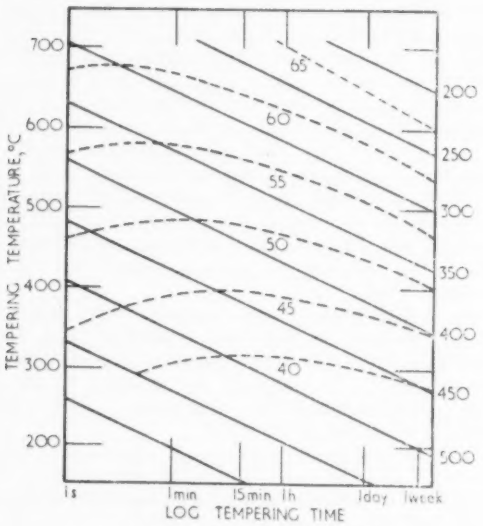

Fig. 5—Tempering of steel *B*, hardness *v.* ductility

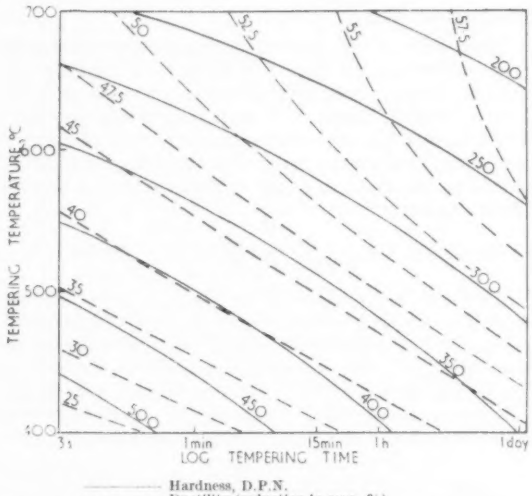

Fig. 6—Tempering of steel *C* (0·5%C), hardness *v.* ductility

a series of X-ray diffraction experiments was carried out.

It would have been interesting to follow the change in the diffraction lines from the carbides, as tempering proceeded, but because of the difficulty in doing so, attention was confined to the ferrite lines.

Two series of specimens, all of plain carbon steel *C*, were hardened and tempered. Those in one series were tempered for 26·8 h at various temperatures, and those in the other series for 5 s at various temperatures. Specimens for X-ray examination were polished and heavily etched to remove cold-worked metal. They were then illuminated with cobalt radiation in a back-reflection camera. The 310 line was photographed and photometered to give an estimate of line-width.

The results are shown in Fig. 8, where both series are plotted on one graph of hardness/line-breadth. With the possible exception of the hardest specimen, it appears that if two hardened and tempered specimens of the same steel have the same hardness, they

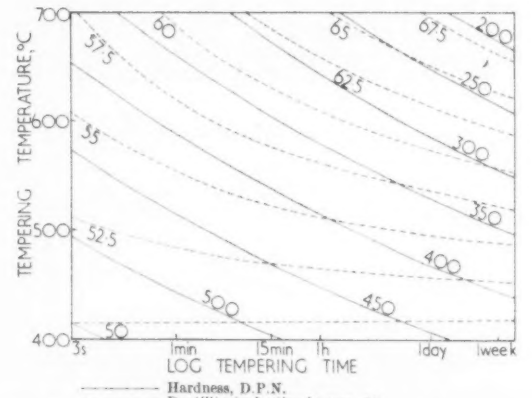

Fig. 7—Tempering of steel *D* (0·5%C, 1·0%Cr), hardness *v.* ductility

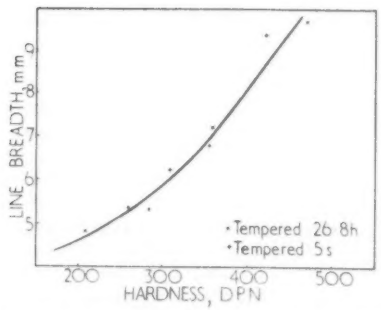

Fig. 8—X-ray line breadth *v.* hardness in tempered steel

will also have the same ferrite line breadth, irrespective of the tempering conditions used to obtain the hardness. The ferrite line-breadth is an index of ferrite strain.

NOTCHED-BAR TESTS WITH INDUCTION HEATING

Tests were made on a few bars tempered with commercial induction-heating plant, to ensure that the results were strictly comparable with those obtained by the direct-coupled resistance-heating method. This also made it possible to treat bar of sufficient diameter for notched-bar test pieces.

Heat treatment

Twelve bars of 1%Cr steel *D*, 9 in. × ½ in., were hardened as before. Five were tempered at various temperatures for 1 h and quenched in water.

The other bars were induction tempered in industrial plant.* Power was supplied by a motor generator at a frequency of 9·7 kc/s. The heating coil, which was horizontal, consisted of 36 turns of copper tubing, the turns being slightly more widely spaced at the centre than at the ends. The specimen was supported in this coil in a silica tube, and adjusted to be as nearly co-axial as possible. It was heated for 10 s, then after a 5 s delay it was ejected from the tube and fell into a water tank. Different tempering temperatures were obtained by changing the power supply to the coil. Hardness traverses showed that, except at the extreme ends, the temperature had been very uniform throughout the bar.

* Made available by Messrs Birlec Ltd.

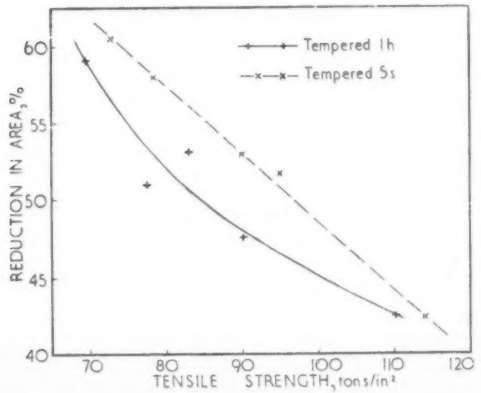

Fig. 9—Tempering of steel *D*, strength *v.* ductility

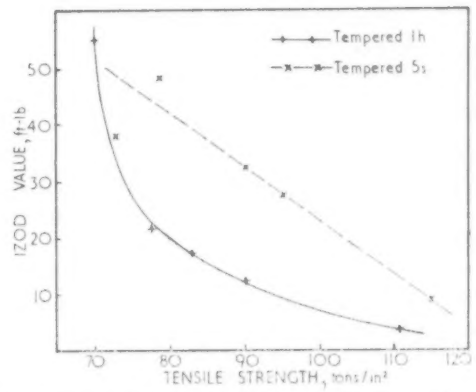

Fig. 10—Tempering of steel *D*, strength *v.* toughness

Results obtained

A 3-notch Izod test-piece and a 0·357 in. dia. tensile test-piece were machined from each of the tempered bars. The mechanical test results are shown in Figs. 9 and 10. These results correspond with those obtained on the small test-pieces, i.e. for a given strength a larger reduction in area is obtained in a specimen tempered for a short time than in one tempered to the same hardness in a longer time and at a lower temperature. As was to be expected, this is most marked in the notched bar results.

MICROSCOPICAL EXAMINATION FOR TEMPER BRITTLENESS

These results, especially from the Izod tests, suggested that the differences in ductility might be due to temper brittleness. Transition-temperature determinations were notp ossible, but microscopical examinations were made, using aqueous picric-acid etch. This etchant makes a well-marked grain-boundary attack in temper embrittled steel,[4] although it does not follow that material which shows this attack will necessarily show a decreased notched bar strength at room temperature.

Samples of two of the Izod test-pieces of steel *D* were mounted, polished, and etched together. One of the specimens had been furnace tempered, and the other induction tempered. They were chosen because they had almost the same hardness, but a marked difference in Izod value. Microexamination showed a clear difference in the degree of grain boundary attack. The improvement in ductility obtained by the use of a short tempering time is therefore due to the avoidance of temper brittleness, at least in this steel.

Microspecimens of the other steels were examined again, using the aqueous picric-acid etch. The 1%Mn steel *B* showed a difference between the specimens tempered for a short time, and those tempered for a long time, but it was less marked than with steel *D*. In the other two steels used, no difference in etching characteristics could be recognized with certainty, so it may be that the differences in properties in the 3%Ni steel are due to some other cause.

DISCUSSION OF RESULTS

Hollomon and others* have shown that the strength properties of tempered martensite depend only on the 'degree of temper,' so that they are independent of whether this degree of temper, denoted by a parameter P, is obtained by long tempering at a low temperature, or by a shorter temper at a higher temperature. The strength of any particular steel is determined by P, although it is not, in general, a linear function of it. P is given by

$$P = T(a + b \log t)$$

where T = tempering temperature, t = tempering time, a and b are constants for the steel.

This has been confirmed by Fisher and MacGregor[2] for a plain carbon steel tempered for times ranging from 6 min to 10 h.

In the work reported here, it has been possible to temper uniformly for shorter periods than have generally been achieved before, when rapid heating has been obtained by dipping into lead baths. Equations for the strength properties similar to that of Hollomon but with at least one more term, are obeyed over the wide range of tempering times used in this work. The equations are inevitably more complex than that of Hollomon because they use the experimental strength value. His use of a 'tempering parameter' enables him to link parameter and experimental value by a curve which takes account of the empirical nature of the strength scale employed.

It is interesting to speculate on the result of extrapolating the hardness values along the time scale. For each steel, if tempering temperature is held constant, hardness is very nearly directly proportional to the logarithm of the tempering time. It is possible to extrapolate back to the as-quenched hardness, and so to determine when tempering started, assuming that the observed logarithmic softening proceeded uniformly. This is equivalent to determining the induction time for tempering. For each individual steel, except plain carbon steel C, the hardness/log time relationship for each temperature would reach the as-quenched hardness at about the same time. Although it is about the same time for each individual steel, it varies between the steels, but only from about 10^{-7} to 10^{-10} s. This suggests that the induction period is of this order, and that the softening then proceeds logarithmically at a rate depending on the temperature.

The results for steel C show the softening taking place more rapidly at the lower temperatures. This suggests that this steel has a substantially longer induction period at lower temperatures, but no reason is offered for the difference in softening rate.

The ductility decreased with increase in the temperature and logarithm of time of tempering, but the relationship was not linear. The combination of the ductility and strength results shows clearly that the simple idea of a tempering parameter is not sufficient for these steels, i.e. it is not necessarily true that two specimens of the same strength will show the same ductility. If they did, then the two sets of curves in Figs. 5, 6, and 7 would locally lie parallel to each other.

This difference in the ductility of specimens of the same hardness is to be explained, partly at least, in terms of temper brittleness. In a temperature range about 500° C, although increased tempering would normally induce increased ductility, if the increased tempering is brought about by increased time of tempering (as opposed to increased temperature of tempering) temper brittleness may increase, and offset the expected increase in ductility.[5] That this did take place in steels D and B is shown by the Izod test results and the microexamination. The development of temper brittleness is a time-temperature process, so a short tempering time will minimize embrittlement.

* Ref. 1 and elsewhere.

CONCLUSIONS

The following conclusions are indicated by this work, for the steels examined:

(1) Degree of tempering, as indicated by a change in hardness or ductility, is nearly:

(i) a linear function of tempering temperature
(ii) a logarithmic function of tempering time.

(2) Although, for a particular steel, a hardness measurement may be used to determine strength properties (and also the width of the X-ray diffraction line produced by the ferrite) it provides a much less reliable indication of the ductility.

(3) In the low-alloy steels examined it was found that the ductility was better if a required hardness were obtained by a short tempering time at a high temperature rather than a longer tempering time at a lower temperature.

(4) As a result of (3), induction tempering is likely to give better properties than furnace tempering, provided that sufficient softening will take place in the time available.

(5) This improvement in ductility by tempering for only a short time can be explained, at least in part, by a reduction of temper brittleness.

ACKNOWLEDGMENTS

The author wishes to express his appreciation of the continuing attitude of interest, encouragement, and trust shown by Prof. J. H. Andrew throughout this work, and to acknowledge many helpful discussions with, and suggestions from, Drs H. Lee, D. V. Wilson, P. E. Brookes, and other colleagues. His thanks are also due to the Department of Scientific and Industrial Research, whose financial assistance made the work possible, to Prof. A. G. Quarrell for his comments on this paper, and to Mr E. F. May and the management of Messrs Birlec Ltd. for providing facilities for induction tempering the larger bars.

REFERENCES

1. J. H. Hollomon and L. D. Jaffe: *Trans AIME*, 1945, vol. 162, pp. 223–249.
2. J. C. Fisher and C. W. MacGregor: *Trans ASM*, 1948, vol. 40, pp. 302–314.
3. H. Krainer: *Stahl Eisen*, 1945, vol. 65, pp. 95–100.
4. B. C. Woodfine: *JISI*, 1953, vol. 173, pp. 240–255.
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Historical note no. 4 (second series)

Nitrogen in early steelmaking

By H. R. Schubert

IT WAS NOT much more than a century ago that the important part played by nitrogen in the metallurgy of iron was theoretically understood.[1] Nitrogen combined with iron forms intensely hard particles of iron nitride which impart great hardness to the surface of steel.

Although not understood in theory, empirical knowledge of the role of nitrogen can be traced back to a very early age; it was practised in the process of cementation or case-hardening known since at least 100 B.C. For cementing the soft iron produced in the primitive bloomery (smelting) hearth to steel, carbonaceous matter such as powdered hoofs and horns, leather, excrements etc. was used in addition to, or even instead of, charcoal. Since they contained nitrogen these were very suitable for obtaining a hard steel. An early example is supplied by a fragment of a scissors blade preserved at the museum at Hanover and dated to the 3rd to 5th centuries B.C. It contains an unusually high percentage of nitrogen.[2]

The first-known specimen among English finds is the cutting edge of a Danish battle-axe of A.D. 895 or 896, found in the river Lea near Stratford in Essex and now in the Pitts-Rivers Museum at Oxford. The edge had been locally carburized by heating in contact with carbonaceous matter and shows a percentage of nitrogen greater than that normally found in bloomery iron.[3] Since the place where it was found is on a route used by Danish invaders, there is little doubt that the axe was imported from Denmark.[4]

The process used in such a case is described in detail in the earliest technical compendium of the Middle Ages called " Schedule diversarum artium," a lengthy treatise on the casting and working of metals and alloys, together with the operations involved in ecclesiastical art. It was written by a German monk named Theophilus who lived, probably in Westphalia, in the 10th century A.D.[5] In the third book (chapter 19) he described case-hardening of soft iron files with the help of organic substances such as the grease of an old hog and leather straps from the skin of a buck-goat tied on to the iron with flaxen thread. For temporary iron instruments (chapter 21) he recommended urine of a three-year-old buck-goat or that of a young red-haired boy. Similar passages are in the so-called *Mappa Clavicula*, but only in its fuller form of the 12th and not in that of the 10th century. For tempering steel to be used for cutting glass, the milk of a Saracen she-goat that had been fed upon ivy and milked by scratching her udders with nettles was recommended. The young red-haired boy of Theophilus is replaced here by a small girl of ruddy complexion.[6]

Case-hardening with organic substances was still used in 17th century England, according to the description supplied by John Moxon, the first English publisher of a series of technical handbooks. Cow horns or hoofs, he wrote in 1677, were first dried and pulverized and then mixed with sea salt and stale urine or wine vinegar. The iron to be cemented was covered all over with the mixture and encased in loam. When the loam cover had been hardened, the encased iron was heated on the hearth of a blacksmith's forge to a bright red heat (about 900° C), and quenched in water at the end of the process.[7] Although according to Moxon, the process was less practised in his days than earlier, it had not ceased completely. In 1694 an English magazine pointed out that " Hoofs are excellent Fuel for Case-hardening of Iron."[8]

A technical encyclopedia of 1704 reported that steel is made " by stratifying of Plates of Iron in a large Furnace with the Horns or Nails of Animals, under which is made a great Fire: Thus is the Iron calcined, but when red hot, and near melting, they take the Plates out and dip them suddenly in cold Water, and they become Steel."[9] In the later part of the 18th century, the view was still held that the renowned swordblades of Damascus were hardened by quenching them red hot in " fresh goat's blood." The process had to be repeated nine times, " which makes their blades so hard as to cut iron."[10]

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[1] J. Percy: "Metallurgy: Iron and Steel," p. 50: 1964, London, John Murray.

[2] E. H. Schulz: *Sudhoffs Archiv für Geschichte der Medizin und Naturwissenschaften*, 1958, vol. 42, pp. 62–64. I am greatly indebted to Professor Schulz for supplying the following analysis: C 0·15%, P 0·12%, S 0·05%, N 0·15%, Cu < 0·1%, Si and Mn 0·001% each; examination of the structure showed distinct nitride needles.

[3] H. H. Coghlan: " Notes on Prehistoric and Early Iron in the Old World," pp. 191–192 (nitrogen 0·0085%), Plate XV, Figs. 1–4, and Fig. 56, no. 3.

[4] H. H. Coghlan: Private communication.

[5] English translation by Robert Hendrie, " An Essay upon various arts," in three volumes: London, 1847; with regard to the date, *see* O. Johannsen, " Geschichte des Eisens," pp. 74 and 197 (reference to Bishop Barnward of Hildesheim, who lived in the 10th century), 3rd ed.: 1953, Düsseldorf, Stahleisen.

[6] L. Thorndike: " A history of magic and experimental science", vol. 1, pp. 767–769: 1929, New York.

[7] H. R. Schubert: " History of the British Iron and Steel Industry, *c.* 450 B.C. to A.D. 1775 ": 1957, London, Routledge and Kegan Paul.

[8] *A Collection for Improvement of Husbandry and Trade*, edited by John Houghton, vol. 5, no. 119, London, 9th November, 1694.

[9] G. Harris: " Lexicon Technicum ": 1704, London.

[10] G. Smith: " The Laboratory; or School of Arts," vol. I, p. 224, 6th ed.: 1799, London (first published in 1738).

N E W S

Announcements and News of Science and Industry

THE IRON AND STEEL INSTITUTE

Autumn General Meeting

The Autumn General Meeting of the Institute will be held in London from Wednesday to Friday, **2nd to 4th December.** The main technical sessions will be devoted to the presentation and discussion of a series of papers on hot and cold deformation, including forging and extrusion. There will also be sessions devoted to the evolution and prevention of fumes in steelmaking processes and to physical metallurgical subjects. Full details will be published in a later issue of the *Journal.*

Determination of Gases in Metals

The Society for Analytical Chemistry, The Iron and Steel Institute, and The Institute of Metals are holding a joint symposium on Tuesday and Wednesday, **3rd and 4th May, 1960,** at Church House, Westminster, on "The Determination of Gases in Metals." Three sessions are envisaged, occupying the afternoon of the 3rd and the morning and afternoon of the 4th. At the first session, invited review papers will be presented, covering the determination of oxygen, hydrogen, and nitrogen in a general way; the other two sessions will be devoted to shorter papers and discussions dealing with specialized aspects of the field.

Preprints of all the papers will be available before the meeting; the proceedings, including discussions, will be published in a single volume later.

Twelfth Hatfield Memorial Lecture

The Twelfth Hatfield Memorial Lecture will be delivered at Sheffield on the evening of Wednesday, **25th November,** by Academician G. Kurdjumov, of the Department of Physical Metallurgy, Central Research Institute for Ferrous Metallurgy, Moscow. Full details will be announced later.

Darby Anniversary Meeting

In 1709 Abraham Darby achieved the first successful use of coke in ironmaking at Coalbrookdale. To celebrate the 250th anniversary of this event, a meeting will be held from Wednesday to Friday, **23rd to 25th September,** at the University of Birmingham and at Coalbrookdale, Shropshire.

Allied Ironfounders Ltd. have made arrangements to preserve the furnace and to establish a museum on the site of Abraham Darby's experiments at Coalbrookdale. A committee set up under the chairmanship of Sir Frederick Scopes is organizing the meeting.

The following is an outline programme of the meeting:

Wednesday, 23rd September

Evening

Opening lecture at the University of Birmingham by Professor T. S. Ashton on "*The economic and social background of the iron industry in the eighteenth century.*"

Thursday, 24th September

All day

Presentation and discussion at the University of Birmingham of the following papers:

"*The charcoal iron industry in Great Britain, with special reference to Shropshire and the development of the industry up to* 1750," by B. L. C. Johnson.

"*The contributions of the Darby family to ironmaking and the iron industry,*" by Dr A. Raistrick.

"*Coalbrookdale* 1709," by Dr R. A. Mott.

"*Ironfounding past and present,*" by Dr J. G. Pearce, C.B.E.

Evening

Dinner at the Grand Hotel, Birmingham.

Friday, 25th September

All day

Visit by motor coach to the Coalbrookdale furnace and museum, the Iron Bridge, and other historical sites in the neighbourhood.

Further details and forms of registration may be obtained from the Secretary, The Iron and Steel Institute.

Memorial to Sidney Gilchrist Thomas

The Newport and District Metallurgical Society has been in existence for 25 years. The Society serves the industries in the Newport and Monmouthshire area, and provides a very useful function in bringing together all those who are interested in technical developments in the iron and steel and non-ferrous industries.

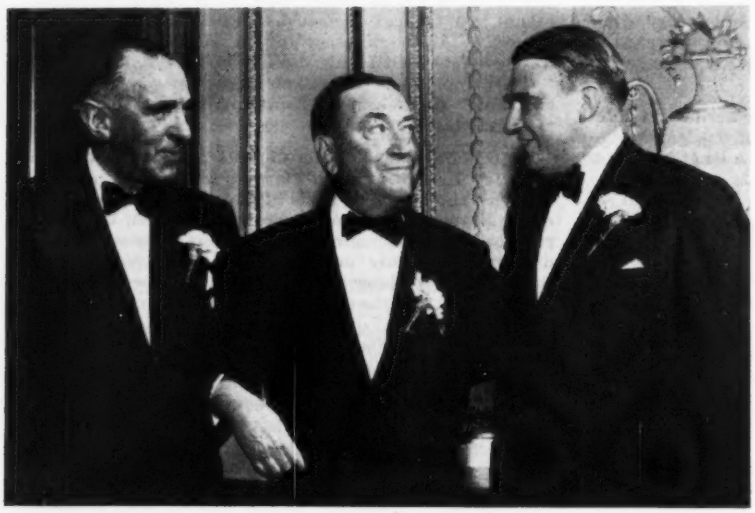
The President and Officers of the Society consider that it is appropriate in their silver jubilee year to honour the memory of Sidney Gilchrist Thomas, who with his cousin Percy Carlyle Gilchrist pioneered the process of basic steelmaking at the works of the old Blaenavon Iron and Steel Company, Monmouthshire. Thomas and Gilchrist carried their early experiments with a basic-lined Bessemer converter for the removal of phosphorus from molten iron, thus making possible, by means of the modern basic Bessemer and open-hearth processes, the exploitation of the vast phosphoric iron-ore deposits throughout the world. Today the majority of the steel produced in the world is made by these two processes.

It is considered that a memorial should be erected on the site of these early experiments to perpetuate their memory. The old Blaenavon Iron and Steel Company's works were closed down recently, but it is fortunate that in their place a modern plant for the manufacture of basic refractories has been erected by Pickford Holland and Co. Ltd. of Sheffield, who owe their very existence to the pioneering work of Thomas and Gilchrist. Mr Ronald Steel, chairman of this company, has generously granted all facilities and permission to erect a memorial on this historic site.

It is proposed to inaugurate a fund for the purpose of:

1. Erecting an obelisk in granite on the site.
2. Forming a trust fund, to be known as "The Sidney Gilchrist Thomas Trust Fund." The income from this fund will be used for:

President and Past-President honoured by AISI



At the 67th General Meeting of the American Iron and Steel Institute, held in New York in May, Honorary Membership of the AISI was conferred upon Mr W. Barr, O.B.E., and Mr C. R. Wheeler, C.B.E., President and immediate Past-President respectively of The Iron and Steel Institute. The presentations were made at the AISI dinner, held in New York on 28th May. The photograph shows Mr Barr and Mr Wheeler at the dinner with Mr Benjamin F. Fairless, President of the AISI and a Bessemer Gold Medallist of The Iron and Steel Institute.

(i) The award annually of a cash prize, to be known as the "Sidney Gilchrist Thomas Prize," to authors of any nationality for the best paper of the year relating to developments in basic steelmaking published in the *Journal of The Iron and Steel Institute*, with which institute the Newport and District Metallurgical Society is affiliated.

(ii) The award of travelling scholarships, if funds permit, to candidates in industry who have completed an advanced course in metallurgy, to study developments in steelmaking practice in another country.

The Council of The Iron and Steel Institute is in sympathy with these proposals, and is prepared to help the Newport and District Metallurgical Society in implementing them. Mr C. R. Wheeler, C.B.E., immediate Past-President of the Institute, in a letter supporting the proposals, wrote:

"*The debt that the iron and steel industries of the world owe to Sidney Gilchrist Thomas is immense. Without his work we should be deprived for all practical purposes of the use of the great resources of phosphoric ores.*

Much of the pioneer work which led to the successful development of the basic processes of steelmaking was carried out by Sidney Gilchrist Thomas at Blaenavon in Monmouthshire. His work is recognised throughout the world but there is no visible memorial. It is fitting that the Newport and District Metallurgical Society, which has its headquarters in Monmouthshire, should take the initiative in establishing a memorial."

The awards will be made by a joint committee to be set up by The Iron and Steel Institute and the Newport and District Metallurgical Society.

It is intended to ensure that this historic site and the annual award of the prize will forever be associated with the honoured memory of Sidney Gilchrist Thomas who, by his genius, perseverance, and skill evolved a process which transformed the iron and steel industry throughout the world.

Since the fundamental work of Thomas is of international importance, the President and Officers of the Newport and District Metallurgical Society have decided to invite all firms connected with basic steelmaking throughout the world to subscribe towards this fund, as they feel sure that they would wish to be associated with the memorial to so outstanding a genius. Subscriptions should be sent to:

G. H. Latham Esq.,
Newport and District Metallurgical Society,
The Whitehead Iron and Steel Co. Ltd.,
Newport, Monmouthshire.

NEWS OF MEMBERS

Mr **W. O. Campbell Adamson** has joined the board of Richard Thomas and Baldwins Ltd.

Mr **L. Badone** has been appointed research fellow at the physical metallurgy division of the Mines Branch, Department of Mines and the Technical Surveys, Ottawa, Canada.

Mr **W. C. E. Bell** was created a Member of the Order of the British Empire (M.B.E.) in the Birthday Honours.

Mr **H. Blincow** has left Hadfields Ltd. to join the Park Gate Iron and Steel Co. Ltd.

Mr **H. Bull** has been awarded the honorary degree of Doctor of Metallurgy by the University of Sheffield.

Mr **W. L. Cave** has joined the board of Richard Thomas and Baldwins Ltd.

Professor **John Chipman** has been elected President of the Metallurgical Society of the American Institute of Mining, Metallurgical, and Petroleum Engineers.

Mr **J. Dearden**, formerly assistant superintendent, has succeeded Mr T. Henry Turner as superintendent of the metallurgy division of British Railways Research Department.

Mr **P. A. Dixon** has relinquished the position of steelworks engineer of the Appleby-Frodingham Steel Company owing to ill-health, but will continue to undertake special duties with the chief engineer.

Mr **W. H. Everard** has been elected vice-president of the Electric Steel Makers' Guild.

Mr **W. F. Gilbertson** has joined the board of Richard Thomas and Baldwins Ltd.

Mr **K. Headlam-Morley**, Secretary of The Iron and Steel Institute, has been elected an Honorary Member of the Staffordshire Iron and Steel Institute.

Mr **A. Howard** has been appointed assistant to Dr L. Reeve, chief metallurgist of the Appleby-Frodingham Steel Company, with the title of research manager.

Dr **B. B. Hundy** has succeeded Mr A. G. Snape as chief research metallurgist at Steel, Peech and Tozer.

Dr **K. J. Irvine** has been appointed deputy research manager of the research and development department of the United Steel Companies Ltd.

Mr **G. L. Jones** has joined the board of Richard Thomas and Baldwins Ltd.

Mr **W. Keen** has been elected secretary-treasurer of the Electric Steel Makers' Guild.

Mr **David Lane**, formerly Secretary of the British Iron and Steel Federation, has joined the Shell International Petroleum Co. Ltd.

Mr **K. F. Lindstrand** has left Jernkontoret to become works manager at Halmstads Järnverks AB.

Mr **John N. Ludwig**, jun., has joined the International Nickel Co. Inc. as supervisor of metallurgical services.

Mr **H. Magnuson** has retired from C. O. Oberg and Cos. AB.

Mr **D. Manterfield** has retired from the position of chief chemist at Steel, Peech and Tozer. He will remain with the company for twelve months in an advisory capacity.

Mr **B. W. H. Marsden** is now steelmaking superintendent of the Algoma Steel Corporation.

Mr **A. Mettam** has been elected president of the Electric Steel Makers' Guild.

Mr **J. Mitchell**, C.B.E., has been appointed a director of Richard Thomas and Baldwins Ltd.

Mr **F. Moore** has succeeded Mr V. Watkins as chief engineer at Steel, Peech and Tozer.

Mr **J. D. Murray** is head of the new applied metallurgy section of the research and development department of the United Steel Companies Ltd.

Professor **N. J. Petch** has been appointed to the chair of metallurgy at King's College, Newcastle-upon-Tyne.

Mr **F. B. Pickering** is head of the new physical metallurgy section of the research and development department of the United Steel Companies Ltd.

Mr **G. L. Richards** obtained the degree of Bachelor of Science last year.

Mr **K. A. Ridal** has been awarded the degree of B.Sc.(Hons.) in physical metallurgy by Birmingham University.

Mr **W. E. Smith** has retired from the position of chief electrical engineer of the Appleby-Frodingham Steel Company.

Mr **A. G. Snape** has retired from the position of chief research metallurgist at Steel, Peech and Tozer. He will remain with the company for twelve months in an advisory capacity.

Dr **James K. Stanley** is now a section manager in technical development at the Crucible Steel Company of America, Pittsburgh.

Professor **F. C. Thompson** is to retire from the chair of metallurgy and directorship of the metallurgical laboratories of the University of Manchester on 30th September.

Mr **T. Henry Turner** has retired from the position of superintendent of the metallurgy division of British Railways Research Department.

Mr **H. W. A. Waring**, C.M.G., has been elected chairman of the Steel Committee of the United Nations Economic Commission for Europe (ECE), in succession to Mr P. van der Rest.

Mr **V. W. Watkins** has retired from the position of chief engineer at Steel, Peech and Tozer.

Mr **H. Williams** is now with the plant applications engineering department of Associated Electrical Industries Ltd., Manchester.

Mr **W. W. Wood** received a knighthood in the Birthday Honours.

Obituary

Mr **Leonard Kingston Everitt**, B.MET., F.I.M. (*elected* 1928), of Sheffield, on 20th June, 1959.

Dr **James Edgar Hurst**, C.B.E., J.P., M.I.MECH.E. (*elected* 1916), of Lichfield, on 20th June, 1959.

Mr **Wilfred Moody** (*Member* 1920–1953), of Manchester, on 7th June, 1959.

Mr **Wilfred Stanley Williams** (*elected* 1942), of Swansea, in February, 1959.

NEWS OF SCIENCE AND INDUSTRY

Symposium on creep-resisting materials

A symposium on problems of development of creep-resisting materials is being held at Marienbad, Czechoslovakia, from 11th to 13th September. The following subjects are to be discussed: structural theory of creep mechanism and the theory of fracture, low-alloy ferritic and modified stainless 12% Cr steels, austenitic creep-resisting steels, high-temperature alloys and special materials, modern methods of development, and methods of testing high-temperature materials.

Full details may be obtained from the Secretariat of the Symposium, c/o Československá Vědecko-Technická Společnost, Příkopy 29, Prague 1, Czechoslovakia.

Fairless Library

The United States Steel Corporation has recently set up a Fairless Library at the Fairless Works of the Corporation. This library will contain many of the awards received by Mr Benjamin F. Fairless during his long career in the iron and steel industry. Notable among these awards is the Bessemer Gold Medal awarded to Mr Fairless by The Iron and Steel Institute in 1951.

Corrosion engineers form association

At a luncheon given under the auspices of *The Corrosion Engineer* on 29th May, Mr H. M. Powell, technical director of Cathodic Corrosion Control Ltd. and chief electrical engineer of the Constructors–John Brown organization, announced the formation of the British Association of Corrosion Engineers.

The objects of the new association will be to promote the dissemination of technical information about corrosion matters and to develop by means of social activities the free interchange of information among members. In due course the association will progress towards the establishment and acceptance of suitable qualifications for corrosion engineers and the promotion of standardization in the terminology and techniques of corrosion control.

Full details and membership forms may be obtained from the Honorary Secretary, British Association of Corrosion Engineers, 97 Old Brompton Road, London S.W.7.

"Revue de Métallurgie"

The *Revue de Métallurgie*, which has been published by the Société Française de Métallurgie for many years, is to appear, starting with the June 1959 issue, in a new form. Two monthly journals will in future be issued in place of one. The first, which retains the old title, will be devoted to applied metallurgy, and in particular to the making and transforming processes for ferrous and non-ferrous metals. The second, to be known as *Les Mémoires Scientifiques de la Revue de Métallurgie*, will be concerned primarily with the scientific and theoretical aspects of metallurgy, like the old *Revue de Métallurgie*, and will consist of two main sections, covering papers and abstracts respectively.

Tinplate plant for Poland

As a result of negotiations with the Polish Government Agency "Centrozap" and following visits to Poland by Head Wrightson's staff, an order has been placed with The

Head Wrightson Machine Co. Ltd. for two hot-dip tinning lines, designed to give an average output of 10 000 tons of tinplate per year. The new plant will treble Poland's present output of tinplate; this increase will be absorbed mainly by the Polish canning industry. Delivery of the equipment will be completed in 12 months.

New laboratories for coil spring research

New laboratories for the Coil Spring Federation Research Organisation at Sheffield, which will be ready in July, will be among the most comprehensive of their kind in the world. They will enable research to be carried out on springs made from the finest wires to bars of up to 2 in. dia. In addition to work on springs and torsion bars, research will be possible on materials in wire and strip form. Metallurgical research will also be undertaken on a very wide range of ferrous and non-ferrous alloys.

A major feature of the new buildings will be a comprehensive electroplating laboratory, which has been presented to the C.S.F.R.O. by W. Canning and Co. Ltd., of Birmingham. The new electroplating laboratory will enable further work to be done in the field of hydrogen embrittlement.

The effect of non-metallic coatings on fatigue life is being studied. So far tensile hardness, abrasion, and corrosion tests have been devised for thin films and carried out on thirteen different resin and plastic protective materials. The research programme for this year also includes work to determine the influence of type of material, heat-treatment, and surface finish on the fatigue life of heavy-duty springs.

Industrial publications received

Bulletin 307 from Acme Conveyors Ltd. deals with the application of controlled vibration to conveyors.

The high-precision cylindrical grinding machine model 1212 is the subject of a new brochure (list 255) from A. A. Jones and Shipman Ltd.

W. C. Holmes and Co. Ltd. have issued a pamphlet describing the Holmes-Schneible SW cupola collector, designed to reduce emissions from cupolas (publication 79).

Publication 46 from Keith Blackman Ltd. is devoted to their Series TF fabric-type dust collectors.

A folder from the Morgan Crucible Co. Ltd. concerns the care and use of crucibles.

Publication No. 321 from Electrothermal Engineering Ltd. describes armoured heaters for preheating and stress-relieving of welds and other forms of heat-treatment. Also of interest is their publication No. 326, which deals with the "Flexible Furnace," a flexible surface heater for temperatures up to 1 000° C.

Recent literature from the Morgan Crucible Co. Ltd. includes brochures on Morganite carbon brushes, Morganite unit-type seals, the Birlec-Morgan electric die-casting furnace (type E.D.F. Mk. II), and the Morgan oil-fired basin tilting furnace (type BT, Mk. II).

Thompson L'Hospied and Co. Ltd. have issued illustrated folders on Duraspun centrifugal castings and heat-resisting castings.

Electric box-type furnaces are the subject of publication no. RF.1/59 from Royce Electric Furnaces Ltd.

The range of British-built Cincinnatti guillotine shears are shown in a new 12-page catalogue from the Cincinnatti Shaper Co. Ltd.

Changes of address

The offices of Dorr-Oliver Co. Ltd., formerly at Abford House, Wilton Road, London S.W.1, and the Mechanical Engineering Division, formerly at 21–22 Manor Mount, London S.E. 23, have moved to Norfolk House, Wellesley Road, Croydon, Surrey (MUNicipal 2488). The laboratory remains at 21–22 Manor Mount, London S.E.23.

CONTRIBUTORS TO THE JOURNAL

J. L. Aston, M.A., PH.D.—Foundry Section Leader, R. and D. Dept., Head, Wrightson and Co. Ltd., Thornaby-on-Tees.

John L. Aston was born at Mirfield, Yorkshire, in 1923 and educated at Bournemouth Grammar School and Bablake School, Coventry. He entered Cambridge University in 1941, and in the following year was directed into industry under wartime regulations. After working on heat-treatment and analytical problems at Morris Motors Ltd., Coventry, and with Wild-Barfield Electric Furnaces Ltd., Watford, for four years, he resumed his studies in the Metallurgy Department at Cambridge. From 1947 to 1950 he was engaged on research into the rapid heat-treatment of steel at Sheffield University. He then spent a sabbatical year at the London Bible College, as a result of which he gained the Diploma in Theology of London University. After a short time at Wednesbury Technical College, he went to Loughborough College of Technology in 1952 to take charge of the teaching of metallurgy to engineering students. He took up his present appointment in 1958 in order to gain further industrial experience.

B. L. Biggs—Research Department, Hadfields Ltd.

Lyndon Biggs was born at Sheffield in 1922. He received his education at Woodlands College, Sheffield. He joined the research staff at Hadfields Ltd. in 1939, and in 1949 took charge of the works observation section of the research department, dealing with works problems connected with metallurgical quality control.

W. A. Tiller, M.A.SC., PH.D.—Westinghouse Electric Corporation, Research Laboratories, Pittsburgh.

William A. Tiller was born in Toronto in 1929. He studied at the University of Toronto, obtaining the degree of B.A.Sc. in engineering physics in 1952 and the degree of M.A.Sc. in metallurgy the following year. He was awarded his doctorate in metallurgy in 1955. Dr Tiller joined Westinghouse in 1955 to continue his thesis work on solidification. He is at present heading a group working to develop a quantitative understanding of all the phenomena associated with freezing.

J. L. Aston

B. L. Biggs

W. A. Tiller

BRITISH IRON AND STEEL INDUSTRY TRANSLATION SERVICE

The following translations are now available, in addition to those given on page 300 of the July, 1959 issue of the Journal:

1176. WINTERKAMP, H.: "Production of Hot Rolled Wide Strip with Particular Reference to the Steckel Mill." *Stahl u. Eisen*, 1954, Nov. 4th, pp. 1492–1502. (£6 5*s*. 0*d*.)

1186. ZINOV'EV, N. V.: "Results of a Discussion on the Quality of Ingot Moulds." *Stal'*, 1958, Nov., pp. 1046–1055. (£6 15*s*. 0*d*.)

1226. WILLMER, T.-K.: "The Principal Impurities in Pig Iron and Their Chemical Determination." *Arch. Eisenhüt.*, 1958, Mar., pp. 159–164. (£3 5*s*. 0*d*.)

1243. POLASEK, J., and E. ZELINKA: "The Planning of Transport in Iron and Steel Works." *Hutnické Listy* (Prague), 1958, July, pp. 626–631. (£5 15*s*. 0*d*.)

1256. WICKE, E., *et al.*: "Reduction of Fine Ores in Two Stages by Carbon Monoxide (CO–C–Iron Process)." *Stahl u. Eisen*, 1959, Feb. 5th, pp. 129–134. (£4 10*s*. 0*d*.)

1277. MALLACH, W.: "The Sulphuric Acid Regeneration Plant in the Acid Circuit of a Wire Pickling Shop." *Draht*, 1958, Sept., pp. 335–342. (£6 10*s*. 0*d*.)

1279. NEUBERGER, F.: "Improving the Wear Resistance of Dies by Treatment with Colloidal Graphite." *Fertigungstechnik*, 1956, July, pp. 321–323. (£3)

1282. GAGIN, S. D.: "The Efficiency of Electric Steel Production by the Duplex Method of Oxygen Converter and Electric Furnace." *Kislorod* (Moscow), 1957, (6), pp. 1–11. (£4)

1292. KULBATSCHNY, I. G.: "Cutting Force and Energy with Parallel Blades" from "Mechanical Equipment for Rolling Mills," pp. 782–788 (a German translation, published by VEB Verlag Technik, Berlin, 1954 of the original Russian book of 1946.) (£2 5*s*. 0*d*.)

1293. *Ibid.* pp. 803–817: "Rotary Shears for Longitudinal Cutting." (£3 5*s*. 0*d*.)

1298. BANDOW, K., and K. WAGNER: "A Contribution on Deoxidation and Desulphurization in the Cupola Furnace." *Giesserei*, 1958, July 17th, pp. 409–417. (£5 5*s*. 0*d*.)

1300. RAUSCH, W.: "The Chemical Surface Treatment of Steel Strip to give a Better Protection against Corrosion, Greater Adhesive Strength of Coatings and Formation of Insulating Layers." *Stahl u. Eisen*, 1959, March 19th, pp. 349–356. (£6)

1304. GEL'MAN, A. S., *et al.*, "The Electroslag Welding of a Hydraulic Turbine Shaft with a Welded Shell." *Svarochnoe Proizvodstvo*, 1958, Sept., pp. 26–36. (£5 15*s*. 0*d*.)

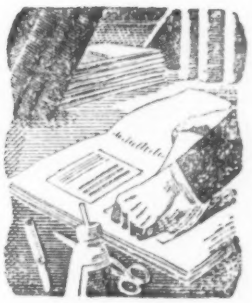
1308. PINSKER, Z. G., and S. V. KAVERIN: "The Electronographic Determination of the Structure of Iron Carbide Fe_3C." *Kristallografiya*, 1956, (1), pp. 66–72. (£4)

1311. SMIRNOV, V. S.: "The Deformation Process in Cross Rolling." *Stal'*, 1947, June, pp. 511–518. (£7)

1351. MISE, S., *et al.*: "Metal Flow in Oblique Rolling." *Sumitomo Metals*, 1957, July, pp. 143–157. (£10)

ABSTRACTS OF CURRENT LITERATURE AND BOOK NOTICES

Iron and Steel Manufacture and Related Subjects



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BLAST-FURNACE PRACTICE AND PRODUCTION OF PIG IRON

Oxygen in Iron Production. R. Durrer. (*Roll'schen Werkzeit.*, 1958, **29**, Nov., 147–149). The function of combined and free oxygen in the reduction of iron ore, and the possibilities of oxygen enrichment in iron and steel manufacture are outlined.

Working Blast Furnaces with a High Top Pressure of 1·5 Atmospheres. L. Ya. Levin, I. A. Kuz'min, V. D. Kailov and A. B. Shur. (*Stal'*, 1958, (11), 964–968). An average increase in output of 1·9% was obtained for every increase of 0·1 atm. in the range 1–1·5 atm. A relative reduction in the coke rate also increased productivity. The main factor in this reduction was the increase of blast temperature to 950–1000°. Variations in furnace conditions over this range of pressures are described and operational factors associated with the top charging gear are reviewed.

The Behaviour of Barium Sulphate in the Blast Furnace. R. Ebert and G. von Struve. (*Neue Hütte*, 1958, **3**, Nov., 676–682). The results of the authors' own experiments and of an evaluation of industrial records are reported. Between 350–600° C $BaSO_4$ is more subject to pulverization than other constituents of the burden. Above 600° C it is reduced to BaS, which passes into the slag, whose viscosity is slightly reduced accordingly. The $BaSO_4$ content of the ore should not exceed the iron content.—L. J. L.

Blast Furnace Operation with a Top-Gas Analysis Meter. II. K. Yasuda. (*Tetsu to Hagane*, 1958, **44**, Mar., 274–276). Continuous CO_2 recording shows conditions of hanging, slipping and channelling. Relationships are shown between CO_2 in top gas and iron analysis, and between furnace conditions, CO_2 in the top gas and iron analysis. There is a good negative correlation between blast volume and CO_2 in the top gas.—K. E. J.

A Model Experiment on Permeability of Blast Furnace Burdens. K. Ono. (*Tetsu to Hagane*, 1958, **44**, Mar., 278–279). Relationships are shown between pressure loss and height of bed, index of flow rate and Reynolds No., and race-way factor and penetration factor on the model furnace.—K. E. J.

On Humidity Control Operation for Blast Furnace. Y. Togino. (*Tetsu to Hagane*, 1958, **44**, Mar., 279–281). Results show that increase in humidity in the blast (in the range 8·8–15·2 g/m³) leads to lower output, increased coke rate and reduced ore/coke ratio.—K. E. J.

Fittings for Blast Furnace Blast Heaters. K. Meyer. (*Demag Nachr.*, 1958, (153), 22–26). The advantages of preheated blast are outlined and a detailed description of the development by Dinglerwerke of blast heating equipment with automatic control is given.

Problems Involved in Heating Materials in Shaft Furnaces with a Layered Arrangement. N. S. Vavilov. (*Doklady A.N.*, 1958, **121**, (5), 845–847). Various methods of heating granular ores or other materials in shaft furnaces are discussed, such as (1) suspended, (2) suspended layer, and (3) suspended fluidized layer. A special shaft furnace containing four diaphragms is described.

The Reduction of Iron Ore Lumps. S. Shono. (*Tetsu to Hagane*, 1958, **44**, Mar., 299–300). Graphs of reduction against time are given for 19 ores, and reducibility index is plotted against indirect reduction ratio for results from two furnaces.—K. E. J.

Utilisation of Manganese in Ferro-Manganese when Operating the Blast Furnace on Oxygen Enriched Blast. A. N. Red'ko. (**Metallurg*, 1958, (**9**), 7–10). A table of operational data is given. The rate of coke combustion depended on burden quality and state of the furnace and was usually 1·6–1·3 t coke per m³ furnace volume per day. The coke charged per round was 2·8–3·2 t; the charging sequence is given. Manganese losses are discussed and slag ratio over the period Mar.–Nov. 1957 is traced.

Industrial and Economical Progress in the Low Shaft Furnace Process at the VEB West Steel Works at CALBE (Saale). W. Schilling and D. Reimann. (*Neue Hütte*, 1958, **3**, Nov., 683–689). The only raw materials used are those that cannot be economically exploited in blast furnaces. Details are given of structural alterations made to the low-shaft furnaces to increase capacity. Measures taken to lower production costs of foundry pig iron are enumerated. The research programme on the process is set out.

Recent Developments of the Hoganas Process. A. C. Lopes. (*Eng. Min. e Met.*, 1958, **28**, Aug., 79–81). [In Portuguese]. The use of sponge iron for steel-making is discussed and details of a modern Hoganas plant are described. Detailed comparative costing is included with a view to assessing the merits of using the process in Brazil.—P. S.

Hot-Metal Pre-Treating Tower. Silicon and Carbon Removal. J. M. Gains and D. C. Hilty. (*Iron Coal Trades Rev.*, 1958, **177**, Nov. 14, 1173–1174). A tower in which pig-iron can be dispersed in droplets and dropped though oxygen is described.

Studies on the Desulphurisation of Molten Pig Iron by Manganese. II. Effect of Si and Mn-S Equilibrium in C-saturated Molten Iron. I. K. Sano and M. Inouye. (*Tetsu to Hagane*, 1958, **44**, Apr., 461–466). Si additions, in the higher range, improve desulphurization by Mn but P has the reverse effect; both effects are negligible below 2%, and hence the desulphurization of B.F. iron depends wholly on carbon saturation. The effects are ascribed to changes in the activity coeff. of S in the melts. The effect of S on the solubility of carbon in Fe–C–Si and Fe–C–P is discussed. (13 references).—K. E. J.

Studies on the Desulphurisation of Molten Pig Iron by Manganese. II. Effect of Si and P on Mn-S Equilibrium in C-saturated Molten Iron—2. K. Sano and M. Inouye. (*Tetsu to Hagane*, 1958, **44**, May, 541–546). Cf. *Ibid.*, 1958, Apr., 461. (13 references).

Measures for Improving the Blast Furnace Gas Take-Off in the Mansfeld Smelting Works "Wilhelm Peck." W. Kurtz. (*Neue Hütte*, 1958, **3**, Dec., 711–720). A description is given of improvements in the design and control equipment of the blast furnace mains and washers of a system supplying blast-furnace gas to a power station.

A Study on the Reaction of Desulphurisation. K. Sawamura. (*Tetsu to Hagane*, 1958, **44**, Mar., 215–217). Changes in sulphur content of pig iron with time at temperatures of 1450°, 1480°, 1500° and 1550° C are shown. The order of the rate-controlling reaction is considered theoretically.—K. E. J.

A Study on Centre-Runner Deoxidation. O. Otu. (*Tetsu to Hagane*, 1958, **44**, Mar., 264–266).—K. E. J.

Investigations of the Equilibrium between Dolomite, Iron (II) Oxide Slags and Iron Melts. W. A. Fischer and H. Spitzer. (*Arch. Eisenh.*, 1958, **29**, Oct., 611–617). The O_2 solubility, equilibrium slag compositions and sulphur distribution was investigated in C-free iron melts in a dolomite crucible, for the temperature range 1530–1700° C.

The Use of Natural Gases in the Reduction of Iron Ores. H. Galluser. (*Blast Furn. Steel Plant*, 1958, **46**, Dec., 1301–1302). Results of trials carried out by Lurgi Ges. für Chemie und Hüttenwesen in conjunction with Ateliers des Charmilles have shown that iron ores can be profitably converted into iron by means of natural gas. An outline of the process and a comparison of costs with blast furnace practice is given.

The Rate of Reduction of Iron Oxide with Hydrogen. L. von Bogdandy and H.-G. Riecke. (*Arch. Eisenh.*, 1958, **29**, Oct., 603–609). Iron oxide pellets were reduced at 600–900° C in a current of H_2. The rate of reduction was examined in relation to temperature and the size of the pellets, and was related to the porosity of the material.

Further Experience with the Kaldo Process. B. Kalling and F. Johanssen. (*JISI*, 1959, **192**, Aug., 330-338). [This issue].

PRODUCTION OF STEEL

Some Consideration on Bessemerizing. Y. Nagano. (*Tetsu to Hagane*, 1958, **44**, Mar., 227–229). Comparisons between the normal and oxygen process are given for decarburizing rate, the relationship between [C]% and (ΣFeO)%, the relationship between bath temp. and refining time and the relationship between time after finish of oxygen injection and Δ[C]%.—K. E. J.

Automatic Control of Basic Converter Heats. K. E. Mayer, H. Knüppel, A. Stumpf and W. Koch. (**Stahl Eisen*, 1958, **78**, Nov. 13, 1670–1676). The authors analyse the individual operations involved in the blowing of basic converters and point out the insufficiencies of the human element in deducing the required information from a great number of individual data in the very short time available for putting decisions into effect, i.e. they stress the advantages of automatic control and discuss the requirements of an automatic plant. Such a plant is being built at the Dortmund-Hörder Hüttenunion at Dortmund. The layout of the plant is described.—T. G.

Gas Flow in the Top-blown Converter. H. A. Hasimoto. (*Tetsu to Hagane*, 1958, **44**, Mar., 222–223). Model studies are reported; the contents of the model converter are mercury and water.—K. E. J.

The Final Composition of Metal and Slag from the Basic Converter. F. Weber, T. Kootz, and K. H. Obst. (**Stahl Eisen*, 1958, **78**, Nov. 27, 1734–1745). Phosphorus content and distribution between metal and slag have been studied for basic converter heats. The studies were extended to oxygen content and to the effect of limestone on both processes. The relationships occurring in the system FeO–CaO–P_2O_5 are pointed out, the effect of particle size of the limestone on manganese slagging is described and the effect of sulphur discussed. The results of the elaborate studies are presented in many graphs and tables and the experience gained in industrial applications of the above results is reported.—T. G.

The Erection and Operation of the New Oxygen-Converter Works in Yawata. T. Uchiyama. (*Tetsu to Hagane*, 1958, **44**, Mar., 224–225).—K. E. J.

The Production of Various Qualities of Steel in Basic Converters with Top Blowing with Oxygen. S. I. Lifshits. (*Stal'*, 1958, (11), 979–983). Rimming steel for deep drawing and killed steel for crane rails equal in mechanical properties to O.H. steel were produced. Further work on sulphur removal is needed before a wider variety of steels can be produced. Steelmaking data are given for the production by top blowing of steel for seamless tubes.—R. S.

Balance of an Experimental Converter on Oxygen Blast. P. Ya. Sorokin. (*Stal'*, 1958, (11), 987–992). Material and heat balances show that about 24% of the heat is lost by volatilization of slag-forming elements. The excess of heat which is characteristic of the process must be used for melting scrap and reducing ore. Dilution of the blast by waste gases or steam to reduce local high temperatures was found to be inadvisable.—R. S.

An Examination of the Simultaneous Formation of Droplet Nuclei and the Growth of Droplets, Illustrated by the Example of the Formation of Dust in the Pneumatic Steel Refining Process. L. von Bogdandy. (**Zeitschrift für Electrochemie*, 1957, **61**, (1), 222–226). An equation for nucleation and droplet growth in condensing vapour is derived, and the diameters are related to carrier gas temperature variation. The basic Bessemer converter is taken as an example and for moderate degrees of supersaturation the equation is shown to apply. Conditions for increasing the size so that removal was greatly facilitated were obtained.

Application of Suspended Steel Form-Panels in the Building of Open-Hearth Furnaces. F. Hammer and A. Czeke. (*Koh. Lapok*, 1958, **13**, July, 336–338). The authors describe the application of suspended steel form-panels in the building of O.H. furnaces, and dicusses the advantages of this method.

Heat Input of the Open Hearth Furnace during Fettling. G. P. Peichev and N. S. Kormilitsyn. (*Stal'*, 1958, (11), 993). At the Kuznetsk Plant great care is taken to control the thermal conditions during all the periods of the heat so that during fettling the heat input in large furnaces is 24 × 10⁶cal/h and in small furnaces 20 × 10⁶cal./h. At Magnitogorsk instructions are to lower by one half or three-quarters the total draught during fettling. This is important for comparatively long fettling periods. Under the best conditions at these plants fettling does not usually last longer than 20–25 min, but in many plants it takes 40–50 min.—R. S.

Contribution to the Study of the Firing of the Open Hearth Furnace. L. Septier. (*Publ. IRSID*, 1958, Series A, (184), Sept. pp. 71). It is shown that a simple calculation can be used to separate the individual factors contributing to the thermal output of the O.H. furnace. The combustion reactions are studied, heat recuperation, heat exchange and methods of firing are discussed and compared.

Grid Gas Firing of Open-Hearth Furnaces. U. Becher and H.-J. Tischendorf. (*Neue Hütte*, 1958, **3**, Dec., 746–752). An account is given of an investigation of the possibilities of using grid gas for firing O.H. furnaces in a plant to be put into operation in Eastern Germany. Firing with a mixture of grid gas from lignite and blast furnace gas was found to be more efficient than producer gas. The low sulphur content of the mixed gas is a further advantage from the point of view of steel quality.

Characteristics and Effect of Additional Tar Oil Firing in 100-ton Open-Hearth Furnaces. F. Schafer. (*Neue Hütte*, 1958, **3**, Nov., 689–694). Theoretical studies of changes in flame characteristics support experiments on the action of additional tar oil firing. Proper application of the practice to large O.H. furnaces may increase output by 10% and more.—L. J. I.

The Oxidation Mechanism in the Open Hearth Furnace. A. L. Hodge. (*Iron Steel Eng.*, 1958, **35**, Nov., 97–100). Oxidation mechanism is discussed, with particular reference to the comparative actions of ore and oxygen, when used in O.H. furnaces working at different rates.

Operation of the New Open Hearth Furnace in Kawasaki Iron Works, Nippon Kokan Ltd. II. J. Shiromatsu. (*Tetsu to Hagane*, 1958, **44**, Mar., 232–233).—K. E. J.

Mixing Steel in Open Hearth Furnaces by Means of Rotating Magnetic Fields. G. A. Ostroumov. (**Stal'*, 1958, (11), 999–1002). A rotating magnetic field such as is used in arc furnaces is suggested. This was used in ingot moulds 20 years ago. Calculations are given showing that the process is quite feasible.

Temperature Control in Open Hearth Furnaces. C. R. Ellstrom. (*Ing. e. Indust.*, 1958, Apr., 89–92). A discussion and comparison of three methods of controlling O.H. roof temperature.

Study on the Reasonable Utilization of Oxygen for O.H. Furnaces. T. Kai. (*Tetsu to Hagane*, 1958, **44**, Mar., 225–227). Graphical relationships are shown between the oxygen rate and the output. Statistical relationships are given between fuel consumption, charge-to-tap time and ore consumption for four levels of oxygen usage (including zero).

Temperature Distribution in Liquid Steel in the Basic Open-Hearth Furnaces. T. Obinata. (*Tetsu to Hagane*, 1958, **44**, Mar., 241–243).—K. E. J.

Some Studies on Automatic Reversal of the Open Hearth Furnace. T. Otsuka. (*Tetsu to Hagane*, 1958, **44**, Mar., 231–232).

Will the Big Demand for Oxygen Come from Openhearths. J. J. Obrzut. (*Iron Age*, 1958, **182**, Sept. 11, 172–174). The effect of such processes as the top-blown converter on the future of the O.H. is discussed. The O.H. is still considered to possess certain inherent advantages particularly when oxygen is used in it.—A. G.

Record-Breaking Open Hearth Furnace. (*Brit. Steel.*, 1958, **24**, Oct., 316–317). Some details are given of the design, and operation of the Ajax tilting furnace at the Appleby-Frodingham branch of United Steel Companies Ltd., which recently produced 19 263 t of steel during a four-week period.

"Ajax" Open-Hearth Furnace at Appleby-Frodingham. Tilting Unit Redesigned for Oxygen Refining. (*Iron Coal Trades Rev.*, 1958, **177**, Sept. 19, 683–684).

Oxygen Increases Open Hearth Output. (*Metallurgia*, 1958, **58**, Oct., 195–196). A description of the full-scale trials at Scunthorpe of the use of O_2 in the modified Ajax tilting O.H. furnace. Details of the modifications to the furnace structure are given.

Oxygen Lancing Trials on O.H. Furnace. (*Refract. J.*, 1958, **34**, Oct., 467–470). Trials by the Appleby-Frodingham Co. are described and modifications made to a 350-t tilting furnace are discussed with their effects on roof wear, refractories, gas cleaning and instrumentation.

Accelerated Firing Rates in Open Hearth Furnaces. E. T. W. Bailey. (*Iron Steel Eng.*, 1958, **35**, Oct., 124–128). An account is given of steps taken to increase firing rates in an O.H. plant of the Steel Co. of Canada Ltd. These are in chronological order from improved charging capacity to checker, blower and burner improvements, instrumentation and increased fan capacity.

Carbon Black in Fuel Oil for Open Hearth Furnaces. G. W. Teskey, jun. (*Iron Steel Eng.*, 1958, **35**, Oct., 71–77). Trials are described showing the substantial improvement in production obtained by adding carbon black to a fuel/mixture consisting of 50% each of oil and natural gas.

The Use of Radio-Isotopes for Study of the Kinetics of Scrap Melting and Slag Formation in the Scrap-Ore Process. A. I. Osipov, L. A. Shvartsman, V. I. Alekseev, V. F. Surov, M. L. Sazonov, M. T. Bulskii, S. A. Telesov, A. M. Skrebtsov, A. M. Ofengenden, L. G. Goldshtein and F. F. Sviridenko. (**Atomnaya Energiya*, 1957, 1193–1197). Melts in 130- and 350-t O.H. furnaces were examined using ^{60}Co which under furnace conditions does not oxidize, volatilize or pass into the slag. It was added to pig iron in the mixer and by sampling the rate of scrap melting was found and an equation developed. Slag formation was studied using ^{45}Ca and the data were similarly treated and an expression derived for the kinetics of slag production.

Light Weight Ladle Crane Trolleys Allow Larger Open Hearth Heats. F. C. Schoen. (*Iron Steel Eng.*, 1958, **35**, Sept., 107–114). By re-designing the crane trolley of a 150-t. hoist, its weight was reduced sufficiently to permit the weight of metal tapped to increase from 130 to 140-t. The trolley was made up of units consisting of weldments and low-alloy steels. Full mechanical details are given.

Desulphurization of Iron and Steel. S. L. Gertsman. (*Foundry*, 1958, **86**, Aug., 48–53). This is a comprehensive review of the methods used and the principles involved in the desulphurization of iron and steel. For iron, CaO, $CaCN_2$, CaC_2, soda ash or NaOH is used, whereas for steel, Mg or 50 : 50 Al–Mg alloy are employed, the latter by lancing in the furnace using an inert carrier gas and the alloy in fine powder form. The operation takes 5 sec. and sulphur levels as low as 0·010% were obtained.—T. G.

Steel Plant Installs Cupolas. (*Steel*, 1958, **143**, Sept. 29, 94, 97). At the Riverdale Steel Plant two cupolas are being built to provide hot metal for oxygen converters. The cupolas have a hearth diameter of 114 in., automated charging, high top pressure, and operate on hot blast delivered by a split wind system.—A. G.

Hot-Blast Cupola Finds Place in Steelworks. (*J. Metals*, 1958, **10**, Sept., 596–598). A summary of articles which have appeared in the German technical press on the use of the

hot-blast cupola in supplying hot metal to O.H. plants and LD converters, and as a unit for smelting up to 10% iron ore and Krupp-Renn pellets.

Electric Arc Furnaces. C. W. Vokac. (*Ing. e Indust.*, 1958, **26,** Jan.-Feb., 53–55, 72). An account of the method of balancing power loads to several arc furnaces.

New Furnaces Speed Delivery of Super Alloys. W. E. Shissler. (*Iron Age*, 1958, **182,** July 3, 70–71). A system installed by the Baldwin-Lima-Hamilton Corp. Pa. is described.

Oxygen in Electric Furnace Benefits Producer and User. (*Steel*, 1958, **143,** Nov. 24, 108–112). The advantages of oxygen refining over ore refining for electric-furnace heats are pointed out. For the producer, oxygen blowing gives positive decarburization control and saves materials, power and refining time; for the user it gives a cleaner steel with better fluidity.—D. L. C. P.

Supplying and Installing an Arc Melting Furnace in India. L. G. N. Edwards. (*Iron Steel Rev.*, 1958, **2,** July, 131–136). An outline of design and accessories adapted to conditions in India is given.

Lighting Flicker Caused by Electric Arc Furnaces. W. E. Schwabe. (*Iron Steel Eng.*, 1958, **35,** Aug., 93–99; discussion 99–100). Erratic changes of arc length occur during meltdown due to collapsing scrap, and "cyclic flicker" occurs due to dislocation of the arc in conjunction with hydrodynamic conditions of the melt near the arcs. The effect of arc current fluctuations on the line network is described, and the influence of the latter on the performance of light bulbs discussed. The effect of furnace and electrode design on flicker is dealt with.—L. D. H.

Trends in Ingot Buggy Design. H. G. Frostick. (*Iron Steel Eng.*, 1958, **35,** Oct., 78–83; discussion 83–86). Factors influencing the design and possible future development of ingot chariots are discussed.

Some Experiments with Steel Ingot "Hot Tops." M. E. Harnish and M. C. McQuarrie. (*Amer. Ceram. Soc. Bull.*, 1958, Aug., 357–360). Small and large scale tests were made and an expression to be used in hot top design is deduced. Heat loss by radiation amounts to 35% or more and it can be stopped by placing a cover over the hot top. Wall thickness has a minor and porosity a major effect on cooling characteristics. The optimum ratio of height to diameter is not constant but varies with size and shape of the mould for which the hot top is to be designed.

Hot Top Practice at Lukens Steel. R. L. Bunting, jun. (*J. Metals*, 1958, **10,** Aug., 525–526). Lukens Steel use reinforced cement hot tops for all hot-topping where the height of the ingot is more than 12 in. from the mould top. The technique of setting the cement in moulds is described, with details of the method of using the hot tops, and their advantages.

Automatic Pouring Ups Yield. (*Steel*, 1958, **143,** Aug. 4, 104, 106). The system of automatic pouring control designed by the Republic Steel Corporation is described. Strain gauges measure the weight of metal and actuate a hydraulic-electric device for closing the stopper.—A. G.

Casting Pit Practice. G. R. Bashforth. (*Sheet Metal Ind.*, 1958, **35,** Oct., 782–791). This is the first of a series on the correlation of theory and practice in rolling. It considers the casting of steel, ingot solidification, cracks, piping, segregation, blowholes surface defects and inclusions and attempts to relate these to turning practice and other operations.

Flow Patterns in Regenerators. M. Simada. (*Tetsu to Hagane*, 1958, **44,** Mar., 229–231). Models are used to explore flow patterns, and to determine the relationship between flow distribution and hydraulic resistance as the position of the checkerwork and the area of checker opening are varied.—K. E. J.

Behaviour of Gases and Non-metallic Inclusions in Liquid Steel during the Basic Open Hearth Process. M. Nakamura. (*Tetsu to Hagane*, 1958, **44,** Mar., 233–235). For various grades of steel (analyses given), relationships are shown between carbon and oxygen content at the refining period, oxygen and the rate of carbon removal, oxygen contents before Mn addition and tapping and the percentage of rejected ingots, increase of N_2 content at tapping and the percentage rejected, and the non-metallic inclusions and the stage after tapping.—K. E. J.

Studies on the Kinetics of Steel-making Reactions. On the Process of Growth of CO Bubbles in Liquid Iron. K. Niwa. (*Tetsu to Hagane*, 1958, **44,** Mar., 221–222).—K. E. J.

Study on the Sampling Method for Molten Steel in Oxygen Analysis. II. Distribution of Oxygen in the Open-Hearth Furnace Bath. Y. Nokagawa. (*Tetsu to Hagane*, 1958, **44,** Mar., 251–253). Results are given for the variation in oxygen content along and down the bath during the oxidizing and refining periods.—K. E. J.

Austenitic Grain-size Control of Medium-carbon and Carburizing Steels. B. L. Biggs. (*JISI*, 1959, **192,** Aug., 361–377). [This issue].

Heat Resistant Loader Banks Furnace Door. (*Iron Steel Eng.*, 1958, **35,** Nov., 142). A tractor shovel used by Granite City Steel Co. to bank the doors of its O.H. furnaces with dolomite is designed to operate rapidly enough to withstand the high temperatures involved.

Development of Electric Arc Furnaces Used for Steel Making. L. Szöke. (*Koh. Lapok*, 1958, **13,** Sept., 427–433). The author reviews the development and present state of electric arc furnaces and describes some up-to-date techniques in steelmaking, such as: slag reaction (Perrin and Coli methods), induction stirring, CaC_2 injection, use of oxygen in the refining period, and vacuum casting.—P. K.

Fundamental Investigation of Bed-Linings of Basic Héroult Furnaces. T. Nishimura. (*Tetsu to Hagane*, 1958, **44,** Mar., 245–247). Chemical compositions are given for three types of lining block, and micro-structures of portions of these after service are discussed. The chemical compositions of slag materials and magnesia and chrome bricks are also given.—K. E. J.

Bath Temperature Measurement in the Electric Arc Furnace. D. Yoshinari. (*Tetsu to Hagane*, 1958, **44,** Mar., 243–245). Thermal e.m.f. characteristics are given for Pt–6%Rh/Pt–30%Rh, Ir — Ir/60%Rh, Pt/20%Rh — Pt/40%Rh, Pr 13 and W–Mo couples, and compared. Typical bath temp. measurements are tabulated.—K. E. J.

Steel Making in the Arc Furnace Without Using Scrap. G. Coll Gomez-Trevijano. (*Inst. Hierro Acero*, 1958, **11,** July-Sept., 168–176). [In Spanish]. Details of the trials are given; the 2 t basic-lined furnace at Araya has now been in production, using 100% pig iron charges, since Feb. 1958. CaO is charged first, the pig is melted down with maximum transformer potential and the melt worked with ore or iron oxide. The steel yield is always greater than the pig iron charged. Economic and technical factors are explained.—P. S.

The Use of Argon when Producing Electric Steel. I. P. Zabaluev. (*Stal'*, 1958, (11), 1004). To improve surface and macro-structure of ingots and reduce metal losses, A was introduced into the casting stream, at first directly into the mould through a "spider" 2–4 min before teeming and then through a central gate. Teeming began into generally unlubricated moulds as soon as the neutral atmosphere was created. Ingots from 1·2·85 t were cast from 112 heats with marked improvement in structure and surface.

Items of Discussion on the Quality of Ingot Moulds. N. V. Zinov'ev. (*Stal'*, 1958, (11), 1047–1055). Present standards for "cupola iron" ingot moulds are discussed with a view to introducing changes in structure, composition, armouring, inoculation and service conditions. Standards for moulds of blast-furnace pig iron are being worked out. The use of semi-permanent moulds for casting ingot moulds and effects of C, Si, Mn, S, and P are considered and a table of optimum compositions is given. Recommendations for structure—absence of cementite, pearlite and pearlite-ferrite and fine graphite in moulds that fail by crazing and ferrite-pearlite (50–80% ferrite) and medium and coarse graphite for moulds that fail by cracking should be ensured.

Report of the Ingot Mold Sub-Committee of the Steel-Making Division. (*Tetsu to Hagane*, 1958, **44,** May, 605–610).—K. E. J.

The Increase of Life of Ingot-Moulds. A. Kralik. (*Koh. Lapok*, Öntöde, 1958, **9,** Sept., 205–215). The author investigated the main factors which influence the life of ingot-moulds, such as: construction, use in steel work, and chemical composition. From the results of the investigations the optimum conditions have been determined.—P. K.

Non-Destructive Inspection of Stopper-Head Bricks, Its Results and Some Problems. I. Matsumura. (*Tetsu to Hagane*, 1958, **44,** Mar., 331–333).—K. E. J.

Damage of Stopper-Head Refractories. Y. Isono. (*Tetsu to Hagane*, 1958, **44,** Mar., 333–335).—K. E. J.

Study on Exothermic Compounds. I, II. K. Kakiuti. (*Tetsu to Hagane*, 1958, **44,** Mar., 314–315; 315–316). Details are given of experiments designed to reduce segregation of C, P, S, Mn and Si in billets using four types of exothermic compound (e.g. one containing $CaSi_2$, mill scale and MnO_2).

Study on Exothermic Hot Topping of Ingots. A. Inoue. (*Tetsu to Hagane*, 1958, **44,** Mar., 309–312). Compositions and physical properties of insulating fire bricks and fireclay bricks are compared, and compositions and exothermal qualities of Vallak Cake and Argen are given. Experimental hot-topping results are given for practices based on Vallak Cake, Argen and fireclay bricks.—K. E. J.

Study of an Exothermic Hot Top. The Foseco Feedex. S. Nishigori. (*Tetsu to Hagane*, 1958, **44,** Mar., 312–314).—K. E. J.

High Speed Photography of the Metal Stream. A. G. Il'in and B. Z. Kononov. (*Stal'*, 1958, (11), 994–995). Interesting photographs of the metal stream from nozzle to mould are shown. The camera used is shown, it takes 150–4000 frames/sec giving a retardation of the action of 10 to 500 times when shown.

Hydrodynamics of Molten Steel at Teeming or Tapping. II. Y. Koike. (*Tetsu to Hagane*, 1958, **44,** Mar., 260–262). Results are given for the effects of length and shape of nozzle and of increase in diameter at the nozzle outlet on the "safety-limit of stopper lift," viz. the minimum lift where dropwise flow changes to laminar flow at constant head.

Recent Views on Solidification Processes in Steel Ingots and on Segregation. B. Matuschka. (*Berg-Hütten Monatsh.*, 1958, Jan., 1–12). The author discusses cooling, solidification and segregation in vertical and horizontal directions. Shrinkage depends on composition, wall thickness, external influences and superheating. Suitable solidification is achieved by rapid cooling, starting from the bottom of the ingot, slow cooling of the head, suitable ratio of diameter to length and by spontaneous crystallization.—R. F.

Improvement in the Production of Ball Bearing Steel at Dneprospetstal. I. P. Zabaluev. (*Stal'*, 1958, (11), 1003). The metal is now heated towards the end of the boil, the temperature being raised 20–30° above tapping temperature and refined under a carbide slag while power is gradually reduced to the furnace. 40–60 min are needed to form the slag and to hold the metal under it, the slag is then removed and the heat tapped under a white slag. Al (0·4 kg) is added 2–3 min before tapping. Refining lasts 2–2½ h, but nevertheless oxide and sulphide inclusions are reduced. During the boil, not less than 0·25%C is removed without overheating and before flushing the oxidizing slag the metal should be at tapping temperature. Tapping under a high viscosity slag gives less inclusions than under a fluid slag of similar composition.

Solute Segregation during Ingot Solidification. W. A. Tiller. (*JISI*, 1959, **192,** Aug., 338–350). [This issue].

FOUNDRY PRACTICE

George Fischer Automated Iron Foundry. I. Mechanized Moulding Matches Manual Skill. (*Metalw. Prod.*, 1958, **102,** Oct. 3, 1727–1732). Mould-making procedure at the Schaffhausen

foundry is described. **Automatic Casting Cuts Costs.** (Oct. 10, 1771–1775). Pouring cooling and shake-out, sand handling and the relation of design to production capacity are briefly discussed.

Simplex Electric Company's Foundry in South Africa. F. D. Nash. (*Found. Trade J.*, 1958, **105**, Aug. 21, 221–224). Recent extensions are described.

Mechanization on a Modest Scale. J. Bain. (*Brit. Found.*, 1958, **51**, Aug., 387–392). The modernization of a small jobbing foundry is described with particular emphasis on mechanical handling. Some information is included on the selection and installation of roller conveyors.—T. G.

Swiss Foundry Modernizes Melting Plant. V. Frey. (*Foundry*, 1958, **86**, Sept., 56–61). The modernization programme recently carried out in the melting department of the Zurich foundry of Escher Wyss Ltd. is described. Except for the largest castings, where 2 cold blast cupolas are used, all melting is done in electric furnaces. Details of the plant are given.

The Common Sense of Cupola Operation. E. Dibden. (*Brit. Found.*, 1958, **51**, Aug., 399–401). The principles of design and operation of cupola furnaces are discussed and the reactions in the furnace and their chemistry explained. Air requirements, volume indication, wind-belt pressure, and the preparation of the sand bed, tap hole and slag hole are briefly dealt with.—T. G.

Automation of Cupola Charging and Stock-line Control with Use of Radioisotopes. G. K. Miroshnichenko, A. G. Vasil'ev, V. I. Shcherbakov and D. A. Lur'e. (*Lit. Proiz.*, 1957, (8), 14–15). A γ-ray absorption device using ^{60}Co is described which by means of relays turns on or off the charging equipment.

From Conventional Lined Cupola to Lining-less Hot-Blast Cupola. E. L. W. Löbbecke. (*Brit. Foundryman*, 1958, **51**, July, 331–342). The author describes the experience gained in the development of the lining-less cupola. The lining-less cupola is in fact usually lined with a protective coating of 1 in. thickness, this soon changes into a coating—about ¼–½ in. thick—of slag, iron and oxides and remains unchanged for years. The melting process is, however, entirely different from that in a conventionally lined cupola and requires far greater blast velocities and volumes at the same blast temperature. In addition, the tuyeres have to be tilted downwards. Design of the latest lining-less cupolas is discussed.—T. G.

Preventing Off-Analysis Heats Through Methods Improvement. H. M. Gowing. (*Foundry*, 1958, **86**, Nov., 79–81). The administrative process whereby an improvement of 60% in rejects was secured is described.

Sight-holes of Cupola Tuyeres. (*J. d'Inf. Tech. Ind. Fonderie*, 1958, Aug., 4–10). Designs with two-position and three-position covers are discussed with special reference to the prevention of explosions.

On Conditions in the Well in Cupola Operation. M. Yohda. (*Nippon Kinzoku*, 1958, **22**, Jan., 33–35). The author has observed that during cupola operation the coke bed eventually floats on the molten metal and slag and the chemical reaction is principally one between metal and slag rather than between metal and coke.—J. O. W.

The Effect of Size of Scrap on the Tapping Temperature of a Cupola. N. H. Keyser and W. L. Kann, jun. (*Mod. Castings*, 1958, **34**, Aug., 59–64). Light, heavy and a mixture of equal weights of these were melted in a 54-in. diameter cupola with a stack height of 22 ft and a metal charge of 2 000 lb at constant blast rate. The change from light to heavy scrap lowered tapping temperatures by ~ 150° F with constant coke charge. The amount of coke needed to maintain tapping temperature is shown for these conditions of operation.

The Relationship between Blast Rate, Coke Charge, Melting Rate and Metal Temperature in the Cupola. (*B.C.I.R.A. Bulletin*, 1958, **14**, Sept., 482–491). A review of formulæ and diagrams previously suggested. (13 refs.).

Measurement of Blast in the Cupola and Sampling in the Melting Zone for Gas Analysis. J. Navarro Alcacer and J. A. Andres Jimenez. (*Inst. Hierro Acero*, 1958, **11**, Jan.-Mar., 58–64). [In Spanish]. Details of the techniques adopted for these purposes are given. The cycle of operation is: 1. determine the height of the coke bed and add coke to bring it to the required height; 2. introduce the gas sampler, which has a length equal to the cupola radius; 3. connect the blower and measure the quantity of air; 4. analyse the gas sample for CO, CO_2 and O_2 and 5. prepare the bed of coke ready to recommence the cycle.—P. S.

Cupola Grit and Dust Arresters. G. E. Lunt. (*B.C.I.R.A.J.*, 1958, **7**, Aug., 317–322). An account of available types of wet and dry dust and spark arresters is given. Costs are shown, and water systems, water consumption, corrosion and abrasion problems and settling tanks and circulating pumps are reviewed.

Controlling C, H_2, N, O in Steel Melting. F. E. Van Voris. (*Mod. Castings*, 1958, **34**, Oct., 28–30, 120–123). Foundry practice is discussed and deoxidation and methods of keeping out hydrogen and nitrogen and some of their effects are considered.

Duplexing Pays at Automotive Foundry. H. A. Laforet and F. J. Webbere. (*Mod. Castings*, 1958, **34**, Oct., 37–40). General Motors, Pontiac Division grey iron foundry duplexing practice for cylinder iron is described and advantages are noted.

Mercury-Arc Converters for Induction Heating. T. A. Cosh. (*Found. Trade J.*, 1958, **105**, Sept. 4, 291–293).

Carbon Arc and Induction Melting, A Comparison. J. A. Edwards. (*Iron Steel*, 1958, **31**, Sept., 449–451). Relative merits of carbon arc and induction melting are considered in relation to the design of the part to be cast, relative cost, ease of control and versatility.

Metallurgical Treatment of Cast Iron in the Induction Furnace. B. Marincek. (*Electrowärme*, 1958, **16**, Oct., 354–365). The melting process and type of furnace are outlined and then the effects of bath movement and of melting in air, of refractory design, loss by oxidation, alloying and recarburation during melting and overheating; and treatments by additions of carbon, basic or iron oxide—containing powders and by gas purging are discussed.

Engineering Aspects of a 30-ton Electric Melting Furnace. C. Hides and J. B. Machin. (*E.S.C. News*, 1958, **12**, Autumn, 4–8). A Birlec furnace erected at Openshaw in Sept. is described.

Probe Key Melting Variables—Find New Facts on Consumable Arc Remelting Under Vacuum. P. M. Unterweiser. (*Iron Age*, 1958, **182**, Sept. 11, 165–168). Preliminary results are given of a laboratory investigation of the effects of current arc voltage, stirring coil power, furnace pressure, crucible cooling water temperature, ingot size and electrode type on the surface condition, micro-porosity, grain structure and sulphur and manganese contents.—A. G.

Gray Iron Chilling Practice. V. Pulsifer. (*Foundry*, 1958, **86**, Sept., 66–72). The results of an investigation to provide the information required to obtain a given depth of chilling for a given grey iron melt analysis are reported.

A New Engineering Metal. W. R. Cooper. (*Financial Times*, 1958, Oct. 22, Supp. I). A general account of spheroidal graphite cast iron. **S.G. Iron Develops Its Own Applications.** A. B. Everest and J. D. Berry. (I). General casting and machining properties are indicated. **S.G. Iron's Resistance to Shock and Corrosion.** A. B. Everest and J. D. Berry (II). Toughness and uses where this is important are discussed. **Meeting Special Demand for Rolls and Dies.** (II). **Power Plant Castings for Land, Sea and Air Uses.** (II). Uses in engines are briefly indicated. **Combining the Virtues of Older Materials.** H. Morrogh. (III). Specifications are quoted and the range of physical properties obtainable are shown. **Expanding Market in All Industrial Countries.** (III). Various uses are quoted. **Equipping Foundries to Make S.G. Iron.** W. W. Braidwood. (IV). The use of Mg-Ni alloy is described. **Joint Work on Quality Control and Development.** P. A. Russell. (IV). **Buyers Guide.** (IV).

Commonsense of Cupola Operation. E. Dibden. (*Eng. Foundryman*, 1958, **24**, Aug., 74–76). Alexander Prize Paper. A general discussion on design and operation of a cupola with special reference to South African conditions, e.g. poor quality of the coke.—T. G.

Blowing of the Cupola by Means of Centrifugal Blowers. (*J. d'Inf. Tech. Indus. Fonderie*, 1958, Feb., 21–25). An attempt is made to put the problem of blowing on a more theoretical basis. The static, dynamic and total air pressure are treated theoretically and the loss of the charge is discussed.—R. P.

Operation and Metallurgy of the Basic Lined Cold Blast Cupola. F. Varga. (*Koh. Lapok*, Öntöde, 1958, **9**, Oct.-Nov., 235–242). The author compares the production of ingot-mould material in basic lined cupolas with that in the acid lined cupola. The higher cost of the former is balanced by the improved quality of moulds.—P. K.

The Present Stage in the Evolution of the Hot-Blast Cupola. (*Tecn. Indust.*, 1958, **36**, Aug., 956–957). [In Spanish]. An account of the Strico-Cupodel experimental hot-blast cupola developed at Buderus Werke in Germany is given. The basic aims of using up to 35% steel scrap, a metal temperature of 1420° C., holding the carbon and silicon contents within limits of ± 0·05% and getting a pearlitic structure have been achieved. Operations so far have shown that the most important needs are to maintain blast temperature at about 500° C and slag thickness constant.—P. S.

Changing the Bottom Portions of the Cupola without Dismantling the Furnace. G. Bienwald. (*Giesserei*, 1958, **45**, Nov. 20, 722–723). Several possibilities of replacing the bottom zone of the cupola in changing over from cold to hot blast practice are outlined briefly.

An Improvement to the System of the Self-Mixing (Cupola) Tapping Spout. (*J. d'Inf. Tech. Indus. Fonderie*, 1958, (100), 1–2). In the improvement described, a greater depth of metal is maintained in the spout, and some increase in carbon is obtained.

Cupola Grit and Dust Arresters. G. E. Lunt. (*Foundry Trade J.*, 1958, **105**, Nov. 27, 651–658, 660). A comprehensive review of wet and dry equipment is given with full illustrations.

Central Supervision and Control of the Cupola Operation. A. Rexroth. (*Giesserei*, 1959, **46**, Jan. 1, 8–11). The installation, equipment and operation of a control cabin in a cold-blast cupola plant is described.

Review of Cooling Water Systems for Cupolas. G. Longaretti and C. Pensotti. (*Fonderia Ital.*, 1958, **7**, Oct., 367–386). The paper gives a comprehensive review of cooling systems for hot-blast cupolas. Design and operating features are discussed in some detail and information is given on such items as quantities of water required under different operating conditions, the behaviour of brickwork and furnace life. (35 References).—M. D. J. B.

Accelerated Working of the Basic Héroult Furnace in a Steel Foundry. A. Mal. (*Fond. Belge.*, 1958, **28**, Nov., 332–334). The procedure described works under a single slag in a reducing atmosphere, without oxidation of the phosphorus.

Principles of a Magnetic Method for Controlling Casting Condition and Quality. A. Karamara. (*25th Internat. Foundry Congress*, Liège and Brussels, 1958, 438–457). [In German]. Tests on grey cast iron are described showing that residual stresses and exterior loading as well as alterations in structure all have marked effects on permeability and coercivity.

Nodular Cast Iron. F. Rodriguez-Yufera y Muñoz. (*Anal. Mec. Elect.*, 1958, **35**, July-Aug., 182–207). A review is given of theories concerning the nucleation and development of nodular graphite.

Spheroidal Graphitic Cast Iron—A New Material. R. Werner and C. Czesany. (*Schweisstechn.*, 1958, **12**, Nov., 141–148). After outlining the properties, method of manufacture and uses of nodular cast iron,

the problem of welding this material is discussed, and the conditions under which satisfactory welds can be made are laid down.

Applications and Properties of Nodular Cast Iron. A. D. Main. (*Inst. Brit. Found. Aust.*, 1958, **9**, 89–101). The strength, fatigue and wear resistance of nodular iron are discussed as well as other properties. Welding is briefly dealt with.

Chemical Composition and Graphite Spheroidization of Imported Pig Irons. I. Aoki. (*Tetsu to Hagane*, 1958, **44**, Mar., 361–362). Results are given for Japanese, Australian, Thai (charcoal), U.S., Chilean, Swedish and Brazilian irons.—K. E. J.

Acid-Resistant Silicon Nodular Cast Irons. T. Dumitrescu, V. Medeleanu, M. Nicolaid and I. Dinu. (*Rev. Mét.* (*Bucarest*), 1958, **3**, (2), 19–27). [In English]. Inoculation with Mg facilitates the production of low C material, reduces the gas content of the castings, and improves the appearance of the machined surfaces, and resistance to corrosion, the last-named property extending to a lower Si content than is usually obtained with lamellar graphite.

Research on the Inoculation of Molten Iron. I. Relation between the Quality of Molten Iron and the Inoculation. M. Okamoto and R. Yoda. (*Nippon Kinzoku*, 1957, **21**, Jan., 18–22). Ca–Si is superior to Si as an inoculant in common (low C, low Si) iron, but is inadequate in strongly deoxidized, e.g. Mg-treated, irons. Fine inoculants are easily oxidized and usually are less effective than coarse; however, fine Cr promotes graphitization. Si alloys are effective with N-containing iron. Orders of effectiveness are: for common iron Si–Al–Ti > Si–Al > Ca–Si > Si > Al > Ti; for high-N iron Si–Al > Si–Al–Ti > Al > Si > Ti. **IV. On the Inoculating Effect of Si–Al and Si–Al–Ti Alloys on Spheroidal Graphite Iron.** R. Yoda. (June, 370–374). With iron containing 0·5%Mg alloy, inoculants (0·2%) were effective in the order Si–Al–Ti > Si–Al > Ti > Al > Si. Small amounts of Al or Al–Ti cause a greater shift of eutectic composition towards lower C contents than Si. The order with iron treated (0·5%) with 2%Mg_3N_2 was Si–Al–Ti > Si > Si–Al > P > SiC. With Mg-treated alloys, Al or Ti (as Si–Al or Si–Al–Ti) must be < 1%, otherwise the spheroidal graphite structure is destroyed. (16 references).—K. E. J.

Investigation on Cast Iron Having Refined Graphite Produced by Melting Cast Iron Covered with Slag Containing TiO_2. X. Experimental Investigation on the Behaviour of Titanium Contained in S–H Cast Iron. M. Tsuda. (*Tetsu to Hagane*, 1958, **44**, Mar., 358–359).—K. E. J.

Modification of Cast Iron with Mg and Na Salts. (*Lit. Proizv.*, 1958, (8), 21). A note taken from *Fonderia*, 1957, **6**, (6) on practice at Allis-Chalmers.

Developments in the Production of Malleable Cast Iron. E. Chapó. (*Koh. Lapok-Öntöde*, 1958, **9**, July, 152–163). The author reviews the chemical composition, mechanical properties, meltings, and heat treatment of malleable cast iron in Hungary and elsewhere, and compares the cost of various melting processes.

Growth of Cast Iron in the Austenite Region. K. Nagaoka. (*Tetsu to Hagane*, 1958, **44**, Mar., 362–365).—K. E. J.

The Role of the Metallurgist in the Production and Usage of Austenitic Manganese Steel Castings. B. Needleman. (*J. S. Afr. Inst. Min. Met.*, 1958, **59**, Dec., 229–257). Composition, metallography, heat-treatment, cold-working, effects of Si, S, P, Cr, Ni, Mo, V and Al, welding, machining, and casting are all outlined and structures are illustrated. Good control of casting temperature is essential and hardness is discussed. Grain size is usefully indicated by a sizing figure.

Non-oxidising Casting of Steels for Prevention of Sand Marks (Including Hair Line Cracks). K. Moriwaki. (*Tetsu to Hagane*, 1958, **44**, Mar., 326–329). Comparisons are made between the non-oxidizing and conventional casting processes; sand marks are analysed at several positions in several ingots.

Nodular Cast Iron. Spheroidal Graphite, Its Origin and Development. F. Rodriguez-Yufera Muñoz. (*Inst. Hierro Acero*, 1958, **11**, Jan.-Mar., 1–23). [In Spanish]. Present day theories of the nucleation and growth of nodular graphite in inoculated cast irons are discussed. The author suggests that nucleation originates directly from the sub-microscopic impurities suspended in the liquid. These dislocation-containing mono-crystals, isolated in the liquid by the supersaturated vapour of the inoculant, begin to grow by the Frank-Read mechanism. The nodule develops in the austenite mass which constitutes a dilute solution to which the Frank theory is applicable. (24 references).—P. S.

Spheroidisation of Graphite by Magnesium Additions to Cast Iron. A. Ya. Khrapov. (*Fiz. Met.*, 1958, **6**, (2), 281–288). The relation between the surface tension of the original cast iron and temperature remains linear within a wide range, but after the addition of metallic Mg it increases rapidly and reaches a maximum in 4–5 min only to decrease thereafter. The results of the experiments point to the conclusion that this high surface activity of Mg cast iron purified from surface active impurities (oxygen and sulphur) is a necessary precondition for the spheroidization of graphite.—L. B.

Applications and Properties of Nodular Cast Iron. A. D. Main. (*Castings*, 1958, **4**, Aug., 9, 11, 13, 15, 17, 19, 21). Typical properties are given and casting properties are outlined and examples of castings are quoted to illustrate the properties considered.

Spheroidization of Cold-Worked Pearlite. A. H. Holtzman, J. C. Danko and R. D. Stout. (*Trans. Met. AIME*, 1958, **212**, Aug., 475–476). A metallographic investigation of the deformation characteristics of pearlite showed that spheroidization commences in areas of slip transverse to the lamellæ.

Formation of Ferrite and Pearlite in Cast Iron. G. Ohira and K. Ikawa. (*Mod. Castings*, 1958, **34**, Oct., 60–66). Iron–carbon, Fe–C–Si and charcoal pig iron were studied. Specimens were cooled from the melt to between eutectic and eutectoid temperature and quenched. At the eutectoid temperature pearlite was nucleated by cementite which precipitated on cooling after solidification; ferrite crystallized on precipitated graphite. Pearlite formation was promoted by P or Sn, ferrite by Mn.

Relation between the Change in the Characteristics of Magnesium-treated Spheroidal Graphite Iron Caused by Remelting and the Behaviour of Residual Mg in Remelted Iron (1st Report). Results of Experiments. K. Okabayashi. (*Nippon Kinzoku*, 1958, **22**, Feb., 73–77). Changes in structure and rates of changes, are described with reference to the time and temperature of re-melting. **(2nd Report). Considerations.** (77–80). From a consideration of results reported in an earlier paper, it appears to the authors that the effective state of residual Mg in spheroidal graphite iron may be found as an active state in the spheroidal graphite domain.

History, Manufacture and Properties of Pearlitic Malleable Iron. M. Tilley. (*Foundry*, 1958, **86**, Aug., 64–69). The author reviews briefly history, metallurgy and types, heat-treating methods, including effect of carbon stabilizers and high sulphur content, and properties of pearlitic malleable iron.

Heat-treated Ductile Iron Offers Wide Property Range. C. A. Anderson. (*Iron Age*, 1958, **182**, Nov. 20, 102–104). A brief discussion of grades and the properties obtained by heat-treating.

Blackheart Malleable Foundry. (*Iron Steel*, 1958, **31**, Aug., 391–394). Extensions for Shotton Bros. Ltd., Oldbury, are described.

Pearlitic Malleable Castings. O. K. Hunsaker. (*Prod. Eng.*, 1958, **29**, Sept., D7; from an ASME paper). The cupola charge, casting technique, annealing treatment and properties of pearlitic malleable castings are briefly outlined.

Malleable Iron Castings. A.S.T.M. Committee A-7. (*A.S.T.M. preprint*, 1958, (5), pp. 2). Proceedings only.

Modern Techniques of Production of Heavy Steel Castings. C. J. Dadswell. (*Brit. Found.*, 1958, **51**, Sept., 435–444). Planning, patternmaking, moulding and coremaking, mould and core materials, drying and closing of moulds, casting, feeding and cooling, head removal, heat treatment and rectification and machining and despatch of castings up to 200 t is reviewed in general terms drawing on both European and U.S. experience.

The Use of Exothermic Feeding Compounds in the Production of Steel Castings. D. H. Snelson. (*Brit. Found.*, 1958, **51**, Oct., 486–501). Feeding characteristics of heads lined with a mouldable exothermic compound have been investigated and principles of head design are discussed. Using a classification based on solidification time, relations for minimum head diameter and section size are developed for castings of simple section. Pouring temperature and running distance are discussed and the results for small castings are extended to much larger designs of the same basic section. The most economical design is specified.

Heat- and Corrosion-Resistant Alloy Castings of the U.S.A. C. R. Sutton and G. F. Geiger. (*Nat. Metallurgical Lab., Council of Sci. and Ind. Res., Jamshedpur*, 1958, 458–475). Alloy steels with 8–14%Cr and C, Ni, Mn, Si, Mo or N are used, and the effects of these alloy additions are considered. High-temperature properties are tabulated and notes on testing are added.

Investigation of Some New Zealand Sands for Foundry Moulding. T. E. J. Seahill. (*N.Z.J. Sci.*, 1958, **1**, June, 342–360).

Packing Characteristics of Typical Foundry Sands. L. J. Pedicini. (*Mod. Castings*, 1958, **34**, Aug., 83–86). Dry and rammed densities and green properties are given for 11 sands with and without bentonite additions.

Controlling Casting Quality Through Sand Testing. H. W. Dietert. (*Foundry*, 1958, **86**, Aug., 58–61). The author describes two practical elevated-temperature tests for moulding sands, one for mould hardness and shock-load expansion and the other for the determination of restraining load. Excellent correlation between test data obtained by shock-load expansion test and casting losses was found for grey iron moulding sands, whereas for steel moulding sands the restraining load test is the best to determine whether or not a steel moulding sand will scab.—T. G.

Moulding Characteristics of Oyaria Sand. R. Santok Singh, B. V. Somayajulu and B. R. Nijhawan. (*J. Sci. Ind. Res.*, 1958, **17A**, Nov., 458–460). The sand is unsuitable for steel foundry practice, but will serve for cast iron.

Moulding Sands Bonded with Wet Clays in Comparison with Dry Clays. V. L. Ré. (*Bol. A.B.M.*, 1958, **14**, Jan., 53–62). [In Portuguese]. Results of tests made with wet and dry clays are given and show that compression strength and hardness values are higher when wet clay is used, in green sand and, in some conditions, in dry sand moulds. Mixing times of not less than four min and not greater than twenty min are recommended. A brief discussion of this paper is appended.—P. S.

Standards for Resin-Coated Sand. F. Le Serve. (*Found. Trade J.*, 1958, **105**, Sept. 4, 297–298). A brief discussion of testing and setting standards for the material is presented.

Batch Sand Distribution and Core Processing. E. J. Ross. (*Brit. Found.*, 1958, **51**, Sept., 420–427). A system is described most of the operations being fully automatic. Loading into and unloading of cores from vertical continuous stoves has been accomplished, and a rotating table stove for drying large cores after pasting and blackwashing is provided.

Centrifugal Pumps Prime Preventive Maintenance. R. E. Jacoby. (*Mod. Castings*, 1958, **34**, Sept., 24–25). Waste foundry sand is reclaimed by mixing into a slurry and classified. The design and replacement system of the centrifugal pumps used ensures that maintenance is reduced to a minimum.

On the Mass-Effect of Special Cast Steel. II. A. Suzuki. (*Tetsu to Hagane*, 1958, **44**, Mar., 341–342). For Mn–Mo, high-C, Cr–Mo and low-C, Cr–Mo steels, the properties of speci-

mens from various parts of a casting are compared, and the mass effect owing to solidification and effect of heat treatment on mechanical properties are tabulated.—K. E. J.

High-Alloy Castings: Selection of Alloys for Heat-Resistant Service. R. A. Miller. (*Met. Prog.*, 1958, **74**, Dec., 98–101). A guide is given to the correct selection of an alloy for elevated-temperature service. The alloys range from 10–30%Cr and 0–70%Ni, balance iron plus minor quantities of other alloying elements.—T. G.

The Use of Continuous Mixers in Sand Treating Plants. S. Laczkovics. (*Koh. Lapok-Öntöde*, 1958, **9**, July, 164–174). A comprehensive review is given of the use of of continuous mixers in sand treating plants, and their advantages over intermittent mixers are discussed.—P. K.

The Influence of Heating on the Strength of Waterglass Sand. H. G. Levelink. (*Metalen*, 1958, **13**, 15 Oct., 340–347). Further results are given on the effect of heating on the compressive strength and deformation under pressure of waterglass–sand mixtures.

Elutriator. T. Wilkins. (*Gjuteriet*, 1958, **48**, Oct., 149–153). A description is given of the construction and operation of an apparatus for determining the silt and clay contents of foundry sands. The method, developed at the Foundry Laboratory of Sveriges Mekanförbund, has been approved for publication in "Swedish Standard Methods of Testing Foundry Sands."

Cement Bonded Sands. R. W. Cox and P. G. Griffin. (*Inst. Brit. Found. Aust.*, 1958, **9**, 53–66). The experience of 2 Victoria foundries with the Randupson process is described. The process is described and illustrated and costs are discussed. As a summary the advantages and disadvantages are set out, among the latter are the long hardening time, difficulty in knocking out and relative impossibility of sand reclamation.

The Design of Moulding Boxes. I. Szilágyi. (*Koh. Lapok*, Öntöde, 1958, **9**, Sept., 215–219). Some basic principles of the adequate design, production, and use of moulding boxes are given, necessary to increase their life. The use of a hardened pilot bush with an oval hole, instead of a round pilot hole, is suggested. Proper standardization of the moulding boxes is also advised.—P. K.

Notes on the History of Patternmaking. L. Faulkner. (*Foundry Trade J.*, 1958, **105**, Dec. 25, 771–774). A brief account from Bronze Age times.

Plastic Patterns. J. W. John. (*Inst. Brit. Found. Aust.*, 1958, **9**, 67–88). Development work first with phenolic and then with epoxy resins is described. The process of pattern making with epoxy resins and fibre glass is given and also corebox construction. Costs are briefly discussed and the need for allergy testing is noted.

Foundry Methods Switch Eliminates Hot Tears in Large Castings, Cuts Weld Time. H. Chappie. (*Western Metalw.*, 1958, Oct., 62–63). Changes involving the use of sand that absorbs expansions, multiple gating of correct thickness, external 1 × 6-in. chills, etc., eliminated hot tears and reduced welding time from 50 h to 1 h per casting. Size of casting is 18 ft × 7 ft × 22 in. Optimum carbon content is 0·40%.—R. S.

The Influence of Weather on the Bonding Properties of Self-Drying Moulding Materials. O. Rácz. (*Koh. Lapok*, Öntöde, 1958, **9**, Oct.-Nov., 242–245). The author discusses the effect of temperature and air humidity on the bonding properties of self-drying moulding materials.—P. K.

Criteria for the Economic Design of Foundry Furnaces. R. Baldantoni. (*Colore*, 1958, **24**, Aug., 381–393). Foundry techniques are briefly discussed and a review is made of mould and core preparation, drying and baking. Factors influencing mould and core construction are discussed in some detail. Means of reducing fuel costs are studied and it is shown that there is scope for economy in furnace design and practice. Vertical furnaces with recirculation are claimed to have low fuel consumption, short drying cycles, flexibility and to give a uniformly baked product with a good cohesion and a minimum use of agglomerating compounds.

Heat Expansion Effects in Clay Bonded Sand Moulds. P. W. Goad. (*Inst. Brit. Found. Aust.*, 1958, **9**, 103–123). The "crust-separation" group of casting defects–scabs, rat-tails and buckles—and the testing of sand for susceptibility to them are considered. A crust-separation test is described and shows that it can occur well below the quartz inversion temperature.

Progress in the Field of the Carbon Dioxide Setting Process. W. Schumacher. (*Giesserei*, 1958, **45**, Nov. 20, 713–721). Difficulties connected particularly with the manufacture of thick-walled castings by the CO_2 process, are discussed in the light of a series of trials, and recent progress in the process is outlined.

Casting Weldments in a Petroleum Refinery. J. Bland, C. B. Parrish and R. C. Wheeler. (*Weld. J.*, 1958, **37**, Aug., 789–798). Some typical applications of casting weldments in a petroleum refinery are described and studies of the weldability of low and intermediate Cr–Mo alloy steel castings are reported. Castings for this purpose must be virtually free from casting defects. Furthermore, typical examples of repair and maintenance of casting weldments are discussed.—T. G.

The Economics of Precision Casting. R. J. Nicholas. (*Inst. Brit. Found. Austral. Branch*, 1958, **9**, 11–29). Equipment and operations at the General Motors-Holden's foundry are described. **The Mechanised Foundry of General Motors Holden's Ltd.** A. G. Bourchier. (31–41).

Dimensional Control of Castings. R. Wiest. (*Giesserei*, 1959, **46**, Jan. 1, 11–14). Methods and standards for the control of the dimensions of castings, including pattern standardization, are described.

Alloy Cast-Iron Recuperator Tubes. Á. Mocsy. (*Koh. Lapok-Öntöde*, 1958, **9**, August, 177–183). The author describes Hungarian experiments on the production of 35 in. long tubes for air heating. The tubes are cast from an alloy cast-iron of the following chemical composition: Si 4–6%, Cr 2·5–3·5% and Ti 0·1–0·4%. They have an adequate constancy of volume, and satisfactory mechanical properties, and are resistant to 900–950° C.—P. K.

Centrifugal Casting Applied to Grey Iron. L. McIntosh. (*Inst. Brit. Found. Austral. Branch*, 1958, **9**, 43–51). Casting machines, their operation and mould speeds, and the metallurgical considerations involved are described.

The Iron Foundry of Innocenti Casts Exclusively by the Croning Process. M. Bertacchini. (*Fonderia*, 1958, **7**, Nov., 483–487). The article describes one of the most up-to-date and best equipped iron foundries in the world—The "Lambretta" motor scooter foundry in Milan. The Croning process is exclusively used. The layout of the plant and production techniques are discussed.

Plastics in Foundry Technique. L. Kolb. (*Gieesereitechn.*, 1958, **4**, Nov., 283–285). The various applications of plastics in foundry practice are outlined, with particular reference to East German methods.

Casting Defects and Their Causes. J. Drachmann. (*Gjuteriet*, 1958, **48**, Nov., 175–180). Various types of inclusions in castings are described and their causes analysed.

The Problem of the Formation of Shrinkage Cavities in Cast Iron. L. Sofroni and R. P. Todorov. (*Rev. Mét.* (*Bucarest*), 1958, (3), July/Aug., 41–57). [In Russian]. The formation of shrinkage cavities in some Fe–C–Si alloys is discussed, existing theories criticized, the authors' experimental equipment is described and the results obtained are discussed.

Co-operation and Co-ordination Between Pattern Shop and Foundry. E. H. Beech and N. Bradshaw. (*Brit. Foundryman*, 1958, **51**, Aug., 363–370). Modern trends and developments of pattern making in close co-operation with the foundry to meet requirements of new casting and moulding techniques and to keep in competition with other types of manufacture are described.—T. G.

Moulding with a Down Sand Frame. W. H. Giles. (*Castings*, 1958, **4**, Aug., 27, 29, 31–32). A frame for attachment to a moulding machine and its operation are described.

Pattern Standards for Practical Foundry Usage. E. A. Geary. (*Mod. Castings*, 1958, **34**, Oct., 90–92). Five categories are distinguished according to the number of castings to be made.

Moulding Profit Cycle. J. N. Carter. (*Mod. Castings*, 1958, **34**, Aug., 22–23). High speed mould production by mould shooting, CO_2 gassing and stack moulding at American Foundry & Machine Co., Salt Lake City are illustrated.

Mercury Patterns for Precision Castings. (*Metallurgia*, 1958, **58**, Sept., 130–134). An outline of the Mercast process, and a description of the equipment and operation of the Crawley works.

Improved Facilities for the Production of Castings by the Mercast Process. (*Machinery*, 1958, **93**, Oct. 1, 777–780). The new Crawley works of Sankey-Telcon Ltd. is described.

Epoxy-Resin Patterns. D. B. Orr. (*Brit. Foundryman*, 1958, **51**, July, 353–354). The author briefly reviews the properties and applications of epoxy resins with respect to pattern work. The 'wet lay-up' technique, which is described, gives the utmost accuracy. Permanent moulds with, e.g., glass-cloth backing should have strengthening ribs to give the required rigidity.—T. G.

Epoxy-Resin Pattern plates. R. Taylor. (*Found. Trade J.*, 1958, **105**, Nov. 13, 593–595). The method of construction is shown and advantages over wood are claimed.

Solidification and Risering of Gray Iron Castings. C. M. Adams jun., M. C. Flemings and H. F. Taylor. (*Mod. Castings*, 1958, **34**, Aug., 31–41). An 8-year programme is reported. Fundamental variables for determining shrinkage of grey iron were found. Size and location of voids are referred to various factors and basic principles of risering are discussed. Castings are shown to illustrate these principles.

Mechanized Molding Line is Fast and Flexible. W. G. Gude. (*Foundry*, 1958, **86**, Sept., 62–65). The use of high-speed automatic moulding machines, together with the organization of material handling have made it possible to produce over 200 moulds/h at the works of the Albion Malleable Iron Co., Michigan.

How to Test and Evaluate Core Oil. R. L. Boyes. (*Mod. Castings*, 1958, **34**, Aug., 24–25). A bakeability test and baked tensile strength determination are described.

CO_2 and Why? A. Kirkham. (*Brit. Foundryman*, 1958, **51**, July, 323–330) After a brief historical review of the process, the author reports on the experience gained in his own foundry when he introduced the CO_2 process. The advantages and disadvantages of the process are compared. The main disadvantages—difficulties in black washing and the necessity to dump all the CO_2 sand after use—are far outweighed by the numerous advantages.—T. G.

An Investigation into the Relative Efficiencies of Some "Breakdown" Additives for Use in the CO_2 Process of Bonding Core Sands. H. T. Bidwell. (*Brit. Found.*, 1958, **51**, Aug., 371–373). The additives examined were chalk (2%), coal dust (2%) and pelleted pitch (1%) and these are compared with cores containing no breakdown agents. It was found that collapsability increases with decrease in binder content; that incipient fusion occurring at a high temperature gives an unusable core sand when no agents are added; that chalk is effective only up to 800° C; pitch is not recommended; coal dust is by far the most promising of the additives added. The tests were carried out with emphasis on suitability for copper-base alloys.—T. G.

Casting Iron Patterns in Zircon Sand. C. W. Yaw. (*Foundry*, 1958, **86**, Oct., 74–75). The advantages are outlined of the iron production patterns, as cast in zircon-sand-faced moulds by the General Motors Corp. at Detroit. Machining costs are reduced and the pattern is left with a good wearing surface.

Steel Castings for Radioactive Service. G. Sorkin. (*Foundry*, 1958, **86**, Oct., 71–73). The revised American specification MIL-S-18262A (SHIPS) and its implications for the manufacture of steel castings for service in radioactive systems are discussed.

Cast Iron Pistons. A. A. Arantes, C. Guimarães and L. A. Lacerda Santos. (*Bol. A.B.M.*, 1958, **14**, Jan., 5–18). [In Portuguese]. The problems of manufacturing cast iron pistons are considered. Compositions, properties, moulding, casting and the various casting defects as well as microstructure of the pistons are given. It is concluded that, (*a*) such pistons can be easily produced with satisfactory mechanical properties, (*b*) the chemical composition cannot be rigidly fixed because of variations in section size, (*c*) in spite of the high phosphorus content of native pig iron the microstructures of the pistons compare favourably with those of foreign origin, and (*d*) the low sulphur content, another characteristic of native pig irons, makes lamellar graphitization difficult in the high strength irons.—P. S.

60 000 lb Stainless Steel Paper Mill Roll. J. H. Schaum. (*Mod. Castings*, 1958, **34**, Aug., 26–30). An account of the induction melting, centrifugal casting and machining of large rolls is given.

Some Metallurgical Aspects of the Pontiac V-8 Engine Pearlitic Malleable Iron Crankshaft. K. B. Valentine. (*S.A.E. Trans.*, 1958, **66**, 81–85). Elastic modulus, fatigue strength and tensile properties are considered, also hardenability and wear-resistance.

Two Interesting Castings. N. Charlton and C. J. W. Hayhurst. (*Brit. Found.*, 1958, **51**, Sept., 411–415). The casting of a planing-machine side-upright weighing 27·5 t and of a turbine blade ring with cast-in blades weighing 1·5 t by C. A. Parsons Co. Ltd. are described in detail and illustrated.

High-Strength Cast Steel Chains. M. Flour. (*Found. Trade J.*, 1958, **105**, Oct. 30, 535–537). Use in a French foundry of a process credited to the U.S. is described. Continuous lengths are cast.

Centrifugal Castings. Ferrous Metals. J. L. Everhart. (*Mat. Design Eng.*, 1958, **48**, Aug., 89–93). Methods and materials are outlined. Properties of cast iron and carbon and low-alloy steels are given and compared with static castings and forgings. Parts are shown. Duplex castings are also referred to.

Centrifugal Casting Upgrades Stainless. K. H. Pierce. (*Steel*, 1958, **143**, July 28, 69–70). By close control of variables, centrifugally cast parts of 12% Cr stainless steel have been produced with tensile strengths over 200 000 psi and elongations above 10%.

The Air-Gap in Permanent Mold Casting. J. D. Keller and N. R. Arant. (*Blast Fur. Steel Plant*, 1958, **46**, Sept., 957–965). Measurements are described to determine the width of the air gap formed in chill-casting between the mould and the casting surface, and its growth with time. This was achieved by analysing heat conduction and storage in the mould and temperature changes in the casting, in the manufacture of chill rolls.

Some Notes on the Shell-Moulding Process. T. H. Weaver. (*Brit. Foundryman*, 1958, **51**, July, 343–347). The author lists nine points that have to be considered if shell moulding is to be successful, and draws attention to the limitations of the process. With these points in mind, the process is described and discussed.—T. G.

Construction of Shell Mould Patterns and Core Boxes. W. A. Wright. (*Mod. Castings*, 1958, **34**, Sept., 97–98). The advantages of standardization in the design of shell machines, patterns, core boxes and auxiliary equipment are discussed.

Investment Casters Push Ceramic Shell Use. T. Operhall. (*Foundry*, 1958, **86**, Oct., 68–70). An illustrated article dealing with the Mono-Shell investment casting process developed by the Misco Precision Casting Co., and using monolithic ceramic shell moulds.

Quality Control Tests in the Shell Process. E. W. Jahn. (*Foundry*, 1958, **86**, Oct., 80–81). A standard pattern has been developed for producing shell-moulded test specimens, for use in quality control tests. The physical tests are inexpensive, and give a comprehensive picture of heats when combined with frequent chemical analysis.

Study of High Temperature Properties of Shell Moulds. R. A. Rabe. (*Mod. Castings*, 1958, **34**, Sept., 86–96). A report of an investigation carried out in co-operation with the AFS Shell Moulding Materials Testing Committee 8-N, to determine the causes of shell-mould casting scrap due to shell cracking and shell expansion defects is presented.

Shell Moulded Cast Crankshafts. H. N. Bogart and H. C. Grant. (*S.A.E. Trans.*, 1958, **66**, 32–37). The production of Ford V8 crankshafts and the advantages of the shell process are described.

Shell Molding for Steel Castings. R. G. Powell and H. F. Taylor. (*Mod. Castings*, 1958, **34**, Aug., 65–70). Surface defects can be avoided by the use of chill-type moulds either of forsterite with high conductivity or of limestone–sand mixtures where endothermic dissociation occurs as well as reduction of CO_2 by the resin binder.

Reduction of Silica in Large Shell Moulds. L. H. VanVlack, R. G. Wells and W. B. Pierce. (*Mod. Castings*, 1958, **34**, Sept., 61–67). The nature and extent of the reduction of SiO_2 at the mould-metal interface are discussed for the case of SiO_2 shell moulds, and methods of minimizing the deleterious effects to the casting surface are suggested.

A Study of the Solidification Process in Grey and White Irons. A. G. Fuller and I. C. H. Hughes. (*B.C.I.R.A.J.*, 1958, **7**, Aug., 288–298). The quenching technique and thermal analysis were used to study eutectic solidification of cast 2-in. bars of white and hypereutectic flake graphite irons. The white iron shows progressive growth of columnar crystals from edge to centre, the centre being completely liquid. The grey iron was similar. A simplified pictorial representation is shown.

The Soundness of Grey and Nodular Iron Castings. A. G. Fuller. (*B.C.I.R.A. Bull.*, 1958, **14**, Nov., 569–574). A review of the literature. Effects of composition variables, effects of variations in the solidification sequence of the iron, the influence of the mould and the use of feeders are discussed. (19 refs.)

Some Curious Wasters, their Cause and Cure. Shedrick. (*Iron Steel*, 1958, **31**, Sept., 461). Reasons for a number of unusual wasters in foundry practice and methods of eliminating them are considered.—A. G.

The Effect of Temperature and Atmosphere on the Iron–Silica Interface Reaction. G. A. Colligan, L. H. VanVlack and R. A. Flinn. (*Mod. Castings*, 1958, **34**, Sept., 54–60). Fe–SiO_2 compacts were subjected to different atmospheres in which the degree of oxidation was controlled by varying the CO_2/CO ratio, the specimens being subsequently examined microscopically and by X-ray diffraction. The results were applied to the mechanism of mould attack in this system.

Calculation of the Casting Characteristics of Bar Stock on Cooling Beds and the Suitability of Certain Cooling Bed Designs for Material of Different Compositions and Profiles. P. Grüner. (*Wire*, 1958, June, 19–26; Aug., 19–26). The principal types of cooling beds are described in detail and illustrated, and their cooling conditions are analysed. Results of cooling tests are reported for several designs of mechanical rake and roller-type beds, the tests covering a range of dimensions and materials of bar stock.—J. G. W.

A Shrinkage Test for Cast Iron. A. Wittmoser, H. A. Krall and L. Huetter. (*Foundry*, 1958, **86**, Nov., 74–78). Test specimen design for a universal test is discussed and a standard form is advanced.

Some Physical Factors Influencing Shrinkage and Porosity in Grey-Iron Castings. K. E. L. Nicholas and I. C. H. Hughes. (*Brit. Found.*, 1958, **51**, Sept., 428–435). Mould rigidity, pouring temperature, and inoculation effects on porosity and other defects are examined. Nodular graphite castings tend to be oversize, but the use of CO_2 process rigid moulds and cement sand gives a contraction relative to the original mould and improves the yield of sound castings. High pouring temperatures favour unsoundness in green sand moulds but promote soundness in rigid moulds. Inoculants causing eutectic nucleation promote unsoundness.

Properties of Moulding Materials at High Temperatures. (VI). Mechanism of Scab Formation. K. Futaki and M. Isotani. (*Rep. Gov. Ind. Res. Inst. Nagoya*, 1958, **7**, Oct., 752–758). Measurements of thermal expansion, hot strength and hot deformation of sand moulding materials were made and various kinds of scabs were reproduced which are classified as α-, Rat-tail, buckle and expansion, β-, Wash, Cut and Erosion, γ-, Cut. Formation of these types are correlated with casting conditions and mould properties.

Quality Control of Melting Stock for Investment Founders. L. S. Taylor. (*Found. Trade J.*, 1958, **105**, Oct. 16, 467–470). A survey of factors in the production of steel melting stock such as contaminating elements, N_2 and H_2 content and control, and a note on costs is presented.

Some Developments in the Fettling of Steel Castings. I. J. Thomas. (*Brit. Found.*, 1958, **51**, Sept., 416–420). Removal of risers by the packed-lance technique, the powder-washing process for the removal of running and feeding systems are described in detail. Both depend upon the oxidation of iron powder to raise temperatures and assist fluxing.

Clean Air and the Foundry Industry. H. T. Angus. (*B.C.I.R.A.J.*, 1958, **7**, Oct., 325–329). A review of emissions and effluents and methods of control. **Operation of the Cupola to Minimize Emissions.** F. M. Shaw. (330–334). Avoiding dust bearing materials in the charge reduces emission, other measures are hardly practical involving the use of special cokes. Fume is inseparable from the use of scrap and this can neither be prevented nor collected at present. Smoke from dirty scrap can be burnt in the cupola stack. Oil or gas burners for kindling should be used. **The Hot Blast Cupola: Fume Problems.** T. R. Eaves. (335–337). Dust is no problem, but very fine dust and fume remains a problem. A tall dispersal stack is the only suggestion so far. **The Foundry Air Furnace: Smoke, Grit and Fume Problems.** R. B. Parish. (338–342). The production of cast iron rolls is described with reference to the emission of fumes. Black smoke and dust should give no difficulty. **Furnaces Used in the Malleable Cast Iron Industry and the Clean Air Act.** H. G. Hall. (343–354). Rotary furnaces with special reference to the Sesci furnace and the malleable annealing furnace are described.

Noise Reduction in Fettling Shops. R. H. Brown. (*BSCRAJ., reprint*, 1958, Aug., pp. 5).

HEAT-TREATMENT AND HEAT-TREATMENT FURNACES

Heat Treatment at Darley Dale. (*Met. Treatment*, 1958, **25**, Sept., 387–388). A new plant is described using town's gas and consisting of 3 heat-treatment furnaces, 2 tempering furnaces, a quenching machine, a turntable charging machine and other equipment. **Heat Treatment of Stampings. New Plant at Darley Dale.** (*Metallurgia*, 1958, September, 141–142). A description of the modernization programme carried out in this works. **The Firth-Derihon Stampings Ltd. Progress in Heat Treatment.** (*Brit. Steel.*, 1958, **24**, Sept., 302–303). A new plant at Darley Dale with furnaces operating at 200–750° C with and 750–1250° C without forced circulation and with loads up to 5-t and tempering furnaces working at 200–750° without forced recirculation is described. Instrumentation is outlined. **New Heat-treatment Plant at the Works of Firth-Derihon Stampings.** (*Machinery*, 1958, **93**, Sept., 10, 620–621). An installation fired by town gas of three heat-treating and two tempering furnaces with an oil-quenching machine and a turntable-type charging machine has been built.

Investigating the Susceptibility of Structural Steels to Overheat. M. F. Alekseenko, G. N.

Alekseeva, G. N. Orekhov and L. S. Fedotova. (*Stal', Supplement Metallogr. and Heat Treatmt.*, 1958, 21–30). Four types of this class of steel were investigated and the findings show that these were little affected by overheating. Traces of a coarse crystalline zone in the fracture appear only when heating to 1300°. The standard mechanical characteristics are the same in overheated and normal conditions. The overheating of steel can be detected from a fracture in a test for impact strength at low temperatures (– 70°) or from the tensile breaking strength of notched specimens tested under a bending angle of 8°.—L. H.

Use of Gas in Heat Treating Operations at Chrysler. R. E. Morken. (*Met. Treating*, 1958, **9**, Jul.-Aug., 8, 10, 19, 20, 30, 32). The use of natural gas in carburizing and hardening is briefly described. The use of controlled atmosphere is also dealt with.

Heat Treatment and Cleaning in Wire Production. E. Hague. (*Metallurgia*, 1958, **58**, Aug., 80–83). Special reference is made to procedure in the new Kingfield Works plant, which is illustrated and described.

Wire Heat Treatment and Cleaning Plant at Kingfield Works. E. Hague. (*Wire Ind.*, 1958, **25**, June, 565–567, 588). The newly opened heat-treatment plant here described comprises a spheroidizing furnace, and hardening, tempering and annealing, pit-type furnaces, all electrically heated. The cleaning plant includes an electrode salt bath and electrically heated rod baker. All these installations are described in some detail.

Mesh Belt Furnaces. (*Wild-Barfield J.*, 1958, **6**, Sept., 10–12). A brief review with illustrations.

Electric Furnace Devlopments. P. F. Hancock. (*J. B'ham Met. Soc.*, 1958, **38**, Sept., 89–104). A general review of arc, induction and resistor furnaces for all purposes.

Application and Heat Treatment of Plain and Alloyed Steels. A. G. Gardner. (*Mech. World*, 1958, **138**, Oct., 438–442). Steels are classified by C content and uses are listed. Notes on annealing, stress-relieving, hardening and tempering are added. Alloy steels are then discussed and the effects of Ni, Cr, W, Mo, Co, Al, Si and Mn are briefly considered.

Salt Bath Heat Treatment *v.* Quench and Temper; Standard and Pearlitic Malleable. P. W. Green. (*Mod. Castings*, 1958, **34**, Oct., 41–46). The methods of production of standard and pearlitic grades are outlined and physical properties, microstructures, mechanical properties and hardness values for the two heat treatment methods are compared. The conclusions are of a preliminary nature.

The Processing of Wrought Steel to High Strength. L. H. McCreery. (*S.A.E.J.*, 1958, **66**, Oct., 70–74). The treatment of SAE 4340 steel to 220 000 minimum U.T.S. is described. The choice of steel, its working, machining and heat treatment require special care, and features are outlined.

Investigating the Thermal Treatment of 6KhV2S Steel for Blades and Shears for the Hot Shearing of Blooms. M. A. Tylkin. (*Stal', Supplt. Metallog. and Heat Treatmt.*, 1958, 251–255). Based on the experience of "Azovstal'" and "Magnitogorsk Combine" a treatment for hardening with oil quenching is suggested followed by a tempering with subsequent air cooling.—L. H.

Heat Treatment Line Needs One Operator. F. V. Horak. (*Metalw. Prod.*, 1958, **102**, Dec. 19, 2229–2230). A line comprising a controlled-atmosphere hardening furnace, spray washer, controlled-atmosphere recirculating tempering furnaces and two atmosphere generators with charging and unloading equipment is described, used for gears, rollers shafts and pins up to 6 in. diameter, of cast and alloy steels and carburized parts.

Recent Advances in Control Systems for Continuous Processing Lines. C. B. Brooks and J. T. Bradford. (*Iron Steel Eng.*, 1958, **35**, Oct., 129–135). An important factor in the improved performance of continuous processing lines has been the recent improvement in control systems; some of these are discussed.

Considerations of the Character and Use of the Heat-Treated Quality Steels. F. Muñoz de Corral. (*Tecn. Met.*, 1958, **14**, Jan.-Feb., 10–24; Mar.-Apr., 48–66). [In Spanish]. In the first part, the concept of hardenability, its measurement and criticisms of the methods used, and the mechanism of tempering are discussed. The correlation of the mechanical properties and the difficulties of characterizing a steel by these properties is also discussed together with a review of the properties required for any particular purpose. In the second part, problems of correct heat-treatment are considered, and, in conclusion, the author suggests three things are required: to know the steels well, to choose the one necessary and treat it well, and to know what can be asked of a steel for a particular purpose. Six appendices, concerned with the techniques of hardenability and steel selection, are included. (39 references).—P. S.

Thermal Treatment of Stainless- Dispersion-Hardened Spring Alloys. A. G. Rakhshtadt. (*Stal' Supplt. Metallogr. and Heat Treatmt.*, 1958, 181–228). The chemical composition is given for the 11 alloys investigated and the findings for each alloy are set out. In particular, regarding alloy N36KhTYu used for elastic parts (diaphragms) a hardening temperature from 950° is suggested with quenching in water and then tempering 2 h at 700°. (The composition of this alloy is: 0·05%C; 34·5–36·0%Ni; 12–14%Cr; 2·8–3·2%Ti; 0·9–1·1%Al).—L. H.

Pit-Type Hardening Furnace. (*Iron Steel*, 1958, **31**, Aug., 424). An installation for axle-shafts, which are suspended vertically to minimize distortion, at Standard Motors, is briefly described.

Heat-Treatment Furnace for Large Fabricated Vessels. (*Engineer*, 1958, **206**, Oct. 3, 534–535). A vertical-flame high-speed recirculating furnace holding vessels 20 ft in diameter and 20 ft long and with a range from 200–900° C is described.

How To Harden Steel Rolls. C. J. McCormick. (*Iron Age*, 1958, **182**, Aug. 28, 60–62). The plant and technique used at Bethlehem Steel Co., in making a large variety of hardened steel rolls are described.—D. L. C. P.

Contribution to the Heat Treatment of Steel Castings with 13% Cr. H. Zeuner and V. Heubner. (*Giesserei*, 1958, **45**, Dec. 4, 747–753). The influence of chemical composition on the structure of 13 and 17%Cr steel was examined for various heat treatments with the object of obtaining maximum toughness. The resulting physical properties are tabulated; it was found that the notched-bar impact value of 13%Cr steel could be improved by 3 to 4 mkg/mm² by using lower austenitizing temperatures, oil quenching and coalescing the carbide.

The Casting and Heat-Treating of Steel Gears. J. M. Calkins and V. E. Zang. (*Met. Treat.*, 1958, **9**, Sept.-Oct., 12, 14). The choice of material and type of heat-treatment to be used in the processing of cast-steel gears is described.—T. G.

The Effect of Various Types of Heat Treatment on the Characteristics of a Faulty 13% Chrome Steel. P. Virdis. (*Fondeira Ital.*, 1958, **7**, Dec., 497–500). The author shows that even highly defective chrome steel castings, such as those exhibiting structural defects such as reticular separation of ferrite and carbon, can be greatly improved by heat treatment.—M. D. J. B.

Economic Advantages of Electric Furnaces. W. H. Holcroft. (*Met. Treating*, 1958, **9**, Nov.-Dec., 10–11, 29, 34, 36, 40). Furnaces are classified according to use and ferrous metal heat-treatment and brazing and sintering furnaces are included. Economic factors are evaluated.

Control of Saltbaths, II. O. Schaaber. (*Draht*, 1958, **9**, May, 175–177). This part of the survey is devoted to a discussion of measuring the bath characteristics, notably its heat transfer properties and the resulting cooling rate.—J. G. W.

Production Methods for Heat-Treatment of Steel Components in Salt Baths. (*Current Engineering Practice*, 1958, **1**, Oct., 20–25). Uses of salt baths are enumerated and processes of case-hardening, martempering and austempering and isothermal (cyclic) annealing are described in "Cassel" and "Rapideep" baths.

Continuous Heating of Wire by the Electrical Resistance Method. O. C. Trautman. (*Draht*, 1958, **9**, July, 279–282). The Trauwood process of direct resistance heating of wire is described with particular reference to strand patenting of high-carbon steel wire, alone or in combination with hot-dip galvanizing, and to other continuous heat-treating cycles, with regard to patenting, mechanical properties of drawn wires treated by the Trauwood process are compared with those of conventional patenting.—J. G. W.

Variation in the Dimensions of Ball-Races by Ageing. P. Prókay. (*Koh. Lapok*, 1958, **13**, Sept., 433–438; Oct.-Nov., 501–503). The author has studied the changes in dimensions of ball races after various hardening and tempering processes. He has found that the variations in dimensions depend on the amount of retained austenite. In order to diminish this, and so to decrease the necessary time of tempering, the ball races should be held after quenching for only a short time in the oil, they should then be cooled down in water of a maximum temperatureof 15° C.—P. K.

Cylinder Sleeves Hardened with Negligible Distortion. (*Met. Treat.*, 1958, **9**, Sept.-Oct., 22). The cylinder sleeves are made of cast iron, the bore varying from about 3 to 7 in. The hardening of the sleeves is carried out in Ajax electric salt-bath furnaces, the line consisting of a mechanized austenitizing furnace, oil-quench bath and tempering furnace with hot-water wash tanks in front of and behind the tempering furnace.—T. G.

Study on a Continuous Gas Carburising Furnace. A. Tsutsui, S. Kawasaki, T. Sato and K. Matsui. (*Nippon Kinzoku*, 1957, **21**, June, 388–392). Operating results are given for an imported furnace, and improvements in regularity of structure, effective case, hardness and deformation are reported.—K. E. J.

Case-Hardening of Free-Cutting Steels. A. Schepers and W. Bartholome. (*Stahl Eisen*, 1958, **78**, Nov. 27, 1745–1748). A number of plain carbon steels containing sulphur and/or lead for free-machining were tested for case-hardenability with respect to the manufacturing process; either in the O.H. furnace or the basic converter. No material difference between the two types of steel was found.—T. G.

Special Control Method in Continuous Carburizing of Truck Transmission and Rear Axle Gears. H. C. Harris. (*Ind. Heat.*, 1958, **25**, Oct., 1981–1990, 1996). A fully automatic gas carburizing plant is described in which the carbon potential of the gas is regulated from dew-point measurements.—A. G.

The Possibility of Increasing the Carburization Temperature Inside a Solid Carburizer to 980° C. A. M. Tarasov and M. R. Semenchenko. (*Metall. Obra. Met.*, 1958, (9), 39–42). The Gorky Motor Works raised carburization temperature from 910° to 950° C in 1954. Now the effects of raising it to 980° are examined with regard to grain size, depth of case and productivity. Steels 20 KhGR, 20KhNM, 20Kh and 20 for which compositions and grain sizes are given were used and gear and cam parts were also examined, hardened to depths of 1, 1·4 and 1·6 mm. Deformation was carefully measured. The process has advantages and shows 40–50% greater productivity than treatment at 910°.

The Use of Alcohols With and Without Additions of Nitrogen Compounds for Case-hardening. W. Klempf and A. Stähler. (*V.D.I.Z.*, 1958, **100**, Dec. 21, 1721–1724). Trials of case-hardening by means of a carburizing atmosphere produced by introducing alcohol mixtures into a furnace at cementation temperature, or by using mixtures of alcohol and nitrogen, have shown that the resulting casehardened product is at least as good as, and in some respects better than those obtained by the use of normal case-hardening media.

Practical Experience with the Foil Method for Shop Control of Salt Baths, Based on the Determination of Carbon Content. K. Winterer. (*Durferrit-Hausmitteilungen*, 1951, Dec.,

7–17). Laboratory and works use of the test is described.

Five Years' Observation of Sulfinuz. Y. de Villemeur. (*Revue M Tijds.*, 1958, **4**, (1), 16–24). A description of the results of five years' experience with the "Sulfinuz" surface treatment. The characteristics of the process and its properties are discussed with the aid of photomicrographs of treated mild steels.

Notes on Surface Hardening by Cementation Treatment with Sulphur: Sulf-Inuz. L. Gascuel. (*PACT*, 1958, **12**, Dec., 486–487). The method and its effect on special steels and on cast iron are described.

Reducing Scuffing and Wear of Ferrous Metals. Surface Treatment by Sulfinuz Process. F. D. Waterfall. (*Engineering*, 1959, **187**, Jan. 23, 116–120). The process and the surface it produces are briefly outlined and a large series of Amsler wear tests in comparison with case-hardened with or without phosphating, nitrided, sulphided, phosphated and MoS_2 treated surfaces is reported. Coefficients of friction are given and notes on fatigue and corrosion resistance are added.

Application of Progressive Induction Heating to Forged Roll Hardening Operations. J. Dugan. (*Iron Steel Eng.*, 1958, **35**, Nov., 65–76). Some effects of heat treatment and quenching on the properties of a forged work roll 21¾ in. diam. × 54 in. body length are discussed, and it is shown that the progressive induction method of heating gives improved control of the hardening operation, with reduced internal stresses and improved grain structure. Satisfactory service performance was obtained.

Measurement of Carburizing Efficiency of Salt Baths. O. Schaaber and R. Fischer. (*Härterei-Techn. und Wärmebehandlung.* (Supp. to *Industrie-Blatt*). 1956, **2**, (10), 89–90, 92–96). Foil tests were carried out systematically over a period of 2 months along with cyanide titrations. The foils contained respectively 1·0% and 0·1%C and were carburized for 20 min at 900° C. The bath was in use for relatively small numbers every night and a continuous drop in carbon value was observed (with wide fluctuations). Difficulties with foils are noted and wire mesh was substituted. The nature of carbon value is discussed at length and a "coefficient of carbon transfer" is proposed. Conditions are indicated for the determination.

How to Check Commercial Salt Baths for Their Carbon Content. B. Finnern. (*Härterei-Technik und Wärmebehandlung*, supplement to *Industrie-Blatt*, 1956, **2**, (10), 99–100). The foil test is described and the importance of activity control is emphasized.

Effect of Barium Acetate on the Activity of Solid Carburizing Compounds. L. K. Luksha. (*Metallov. Obra. Met.*, 1958, (7), 50–52). Activity is greatly increased due to acetone production. This enables wood charcoal to be omitted and replaced with a semi-coke. Tests in acetone vapours are suggested.

Recent Development in Industrial Furnaces. Liekmeier. (*Wire*, 1958, June, 40–42). A continuous Furnace for case-hardening is briefly described. Reference is also made to high-vacuum annealing and the use of cooling water for steam-raising.—J. G. W.

Potential of Higher Temperature Carburizing. (*Met. Prog.*, 1958, **74**, Oct., 134–139). Increased oxidation has retarded work with higher temperatures, but progress has now been reported in the U.K. and in Russia. Furnace design, carburizing rates at temperatures up to 2200° F, structures and grain-growth and other possible limiting effects are surveyed.

Gas-Fired Furnace for Economic Carburising. L. A. Lush. (*Gas*, 195, **21**, Sept., 22–25). Furnace design and control are briefly reviewed.

High Temperature Carburizing is Practical. (*Steel*, 1958, **143**, Aug. 4, 98–100). It is claimed that with close control of temperature and atmosphere many commercial carburizing steels could be treated at higher temperatures (up to 2000° F). Test results on 4 steels are presented.—A. G.

Pit Type Carburising Furnaces Installed in 16-acre Plant. (*Gas World*, 1958, **148**, Nov. 15, 101). An installation of 2 furnaces and 2 gas generators at Tannochside near Glasgow is briefly described.

Gas Carburizing and Case Hardenability. T. W. Ruffle. (*Met. Treatment*, 1958, **25**, Oct., 397–403; Nov. 475–481). The steels used, Jominy testing, carburizing procedures, hardening processes and the properties of the treated metals and the isohardness curve are dealt with.

The Pack Carburising and Annealing of 4½%Ni Cr Mo Case Hardening Steel. C. Dawes. (*Metallurgia*, 1958, **58**, July, 3–9; Aug., 69–75). An examination of the results of carburizing a 4½% Ni–Cr–Mo case hardening steel with various compounds led to the conclusion that a 1%Na_2CO_3 or a 2½%$BaCO_3$ compound gave less excess carbide, and are satisfactory agents in the proportion of 9 parts of old to 1 of new. The modification of this practice for plain carbon steels, and the use of energizers are discussed. Factors are discussed and trials are reported. The normal lower critical temperatures of case and core are shown to be 720° and 675° C respectively. The cooling rate influences the retained austenite of the case in hypereutectoid, but has little effect on eutectoid specimens. When it is retained it can be removed by heat-treatment before sub-critical annealing. In the core hardness is not affected by cooling rate and is lowered by multiple tempering more than by an equivalent single treatment. Procedures are recommended.

Developments in Surface Hardening. E. Mitchell. (*Met. Treatment*, 1958, **25**, Oct., 425–430, 433; Nov., 455–463). A lecture. Carburizing methods are surveyed in some detail, as are carbonitriding and flame and induction hardening. Structures and apparatus are shown.

Surface Hardening of Steel. T. V. Cherian. (*Tisso*, 1958, **5**, Oct., 155–165). A review of the effects of case-hardening, nitriding, carbonitriding, chromizing and siliconizing, and flame and induction hardening.

Three Methods of Controlled Flame Hardening by Oxy-Town Gas. (*Gas World*, 1958, **148**, Sept. 20, 56). Recent machines are briefly described.

Hardening of Rail Ends. G. K. Colligan. (*N.Z. Eng.*, 1958, **13**, Sept., 343–349). The flame-hardening process applied to rail ends and the hardness distribution produced are described.

Fast Heat Cycle Toughens Pinions. (*Iron Age*, 1958, **182**, Nov. 20, 110). The tempering of the threaded ends of truck drive pinions by induction heating is described. The cycle is completed in a few min; by using a sufficiently high temperature for a short period the high-hardenability carburizing alloy steel is adequately treated.—D. L. C. P.

Cold-Heat Machine Hardens Gears under Water. (*Metalw. Prod.*, 1958, **102**, Dec. 19, 2215–2216). Induction heating with a specially designed intensifier shaped to the particular tooth profile is used and quenching is carried out by water jets.

Nitriding of High-Strength Magnesium-Treated Cast Iron. Yu. M. Lakhtin and D. S. Pinchuk. (*Metallov. Obra. Met.*, 1958, (7), 39–42). A pearlitic-ferritic Mg-treated iron with 3·05%C (0·61% bound) and 0·63%Mn, 2·41%Si and 0·1%S was used, Mg introduced after tapping and annealed at 950° C. Nitriding was carried out at 500–800° in steps of 50° for 1, 3, 6, 12 and 24 h. Hardnesses are shown, maximum being attained at 650–700° and not at the lower level appropriate to steels containing Al, Cr etc. The time required is 12–24 h and the core is not brittle. Epsilon-phase appears on the surface.

Gas Heated Atmosphere Furnaces. (*Gas*, 1958, **21**, Sept., 28–30). A brief account of carbonitriding at Ford's motor works is given.

Heat Treatment Plant for Crankshafts. (*Engineer*, 1958, **206**, Nov. 21, 808–809). An account of the continuous heat-treating installation of Scottish Stamping and Engineering at Ayr for throughput of 750 shafts daily is given. The plant is gas-fired.

New Roller Hearth Annealing Furnace with Electronic Control. (*Brit. Steel.*, 1958, **24**, Oct., 330–331, 328). The construction and controls of the furnace recently installed by Birlec in the new stainless steel tube mill of Talbot Stead Tube Co., Ltd., are described.

Where Continuous Annealing is Going. N. B. Jones. (*Steel*, 1958, **143**, July 28, 72–75). Current practice at 16 continuous tin-plate annealing lines in North America is reviewed to determine trends. Large capacity furnaces are favoured heated by gas and electricity in combination. Newest installations have final air cooling in place of water quenching.—A. G.

Annealing of Malleable Iron: Effect of Repeated Annealing on Rate of Second Stage Graphitization. J. E. Rehder and J. E. Wilson. (*Mod. Castings*, 1958, **34**, Aug., 49–52). White iron was re-annealed several times and annealability or maximum annealing rate was determined. Time for the first stage of graphitization was measured by quenching from 1750° F and examining for primary carbide. This took 2½ ± ¼ h. Maximum permissible cooling rate was then found and averaged 20° F ± 2° F/h. Surface effects are noted, though the specimens were protected from oxidation almost completely.

J. & L. to Install New Continuous Annealing Line at Aliquippa. (*Iron Steel Eng.*, 1958, **35**, Sept., 211). Plans are announced for the installation of a modern high-speed continuous annealing line in connection with tinplate production at the Aliquippa Works of Jones & Laughlin. A schematic diagram of the proposed plant is reproduced.

Case History on Induction Hardening Large Drive Gear. V. H. Pagano and C. J. Kropf. (*Met. Prog.*, 1958, **74**, Sept., 86–90). The final drive gear of the track sprocket of the M47 and M48 tanks was formerly case-hardened. The steel used was a 3·25–3·75%Ni steel. Induction-hardened gears use boron steel. The operation sequence of the induction-hardening cycle is described. Service tests proved the superiority of the induction-hardened gears.—T. G.

Automatic Fixtures for Induction Heating. C. A. Tudbury. (*Met. Treat.*, 1958, **9**, Sept.-Oct., 8–9, 62–64). The advantage offered by the application of induction heating is the consistency of the results, once the correct conditions are established. The author discusses some of the possibilities and limitations of automatic fixtures for induction heating.

Effect of Induction Hardening on Tensile Strength. K. Yahagi. (*Tetsu to Hagane*, 1958, **44**, Mar., 349–350). The tensile strength of hardened test-pieces shows a min. at a depth of hardening below 1 mm; pieces hardened after sorbite-treatment are stronger than those hardened after annealing. The strength of pieces tempered at 300° C shows a linear increase with depth of hardening. For a depth of 3 mm, the max. strength is obtained by tempering at 250° C for 2 h.

Nitriding of Steel by the Glow Discharge. H. Knüppel, K. Brotzmann and F. Eberhard. (*Stahl Eisen*, 1958, **78**, Dec. 25, 1871–1880). The physical principles of glow discharge are described, particular emphasis being directed to the quenching of arcs. For treating times of about 20 h hardening depth is slightly greater for glow-discharge hardening than for gas nitriding. No effect of the pressure used was found in the range 1–10 mm. Mixtures of nitrogen and hydrogen in ratios from 9 : 1 to 1 : 9 were most suitable. Current density has little effect on nitriding. In addition, cathodic sputtering is a valuable aid for the removal of passivating films e.g. on stainless steels.—T. G.

Structure of Nitrided Heat-Treatable Steels. T. M. Norén and L. Kindbom. (*Stahl Eisen*, 1958, **78**, Dec. 25, 1881–1891). Microstructural studies on heat-treatable steel samples nitrided by glow discharge and by the conventional gas method showed that at the beginning of the treatment glow discharge gives a greater effect than gas nitriding, but there is no difference in hardness or penetration after the normal treatment time. By

X-ray analysis Fe_2N and Fe_4N were detected in the hardened layer. The great hardness of special alloy steel is due to nitriding of the alloying elements, with consequent distortion of the crystal lattice. It was found that Ti contents of about 1·75% greatly affect the hardness of the layer, particularly at elevated temperatures, ~ 800° C.—T. G.

Decarburisation Rate of Mild Steel. Effect of Crucible Materials on Decarburisation Rate. T. Fujii. (*Tetsu to Hagane*, 1958, **44**, Mar., 237–239). Results are given for crucibles made of pure magnesia, electrofused magnesia, sintered magnesia, pure alumina and sintered alumina (analyses given).

Bright Annealing of Steels and Heat-Resisting Alloys in a Reducing Atmosphere Containing Fluoride. P. Galmiche. (*Rev. Mét.*, 1958, **55**, Sept., 840–846). A process for the bright annealing of chromium-bearing alloys in a reducing atmosphere containing fluoride is described, and an application of the process (based on the O.N.E.R.A. process) to the skin annealing of gas turbine blades is discussed.

Scale-Free Annealing in Rolling Mills and Forging Operations. F. Helfert. (*V.D.I.Z.*, 1958, **100**, Dec. 21, 1731–1736). Various methods of heating in controlled atmospheres are discussed, the equilibrium conditions of the relevant reactions are described and some examples given.

Combination Gas-Electric Tower Furnace Used for Continuous Annealing of Tin Plate. (*Ind. Heat.*, 1958, **25**, Aug., 1574, 1576). A continuous protective atmosphere tin plate strip annealing furnace rated at 15 t/h of strip moving 500 ft/min installed by Youngstown Sheet & Tube Co., East Chicago, Indiana is briefly described. Gas is used for heat-up.

Continuous Annealing of Tin Plate. R. A. McClure. (*Blast Furn. Steel Plant*, 1958, **46**, Dec., 1287–1291, 1294). The continuous annealing process for tin plate is described, and compared with box annealing. It is pointed out that operating costs of the latter are less, but several advantages in the use of continuous annealing methods are enumerated.

A Roller Hearth Furnace for Continuous and Batch Annealing of Stainless Steel Tubes. E. A. Hasnip and M. P. Shaw. (*Process Control Autom.*, 1959, **6**, Jan., 8–14). An illustrated account of the Walsall installation of Talbot Stead Tube Co. with gas-firing and controls and conveyors is given.

Rapid Annealing of Cold-Drawn Austenitic Stainless Steel Tubing. H. Saruhashi. (*Tetsu to Hagane*, 1958, **44**, Mar., 343–344).—K. E. J.

Heat Transmission in a Coil Annealing Furnace. K. Tanaka. (*Tetsu to Hagane*, 1958, **44**, Mar., 388–390).—K. E. J.

Study on the Annealing of Steel Sheets for Galvanizing. T. Matumoto. (*Tetsu to Hagane*, 1958, **44**, Mar., 356–357). For specimens of thickness 0·27 and 1·25 mm, results are given for the effect of annealing temp. (max. between 500° and 1000° C) on physical properties, bending, springiness and grain size.—K. E. J.

A Study on High Speed Tools. XX. Water and Air Annealing of High Speed Steels SKH2 and SKH9. H. Hotta. (*Tetsu to Hagane*, 1958, **44**, Mar., 411–412).—K. E. J.

Theoretical and Practical Considerations for Normalizing Cast Irons. G. Cola. (*Fonderia Ital.*, 1958, **7**, Nov., 433–471). An account is given of extensive researches carried out on the normalization of cast irons. Over 50 castings of different sizes and different grey iron compositions were investigated and it was found that 94½% of the castings showed higher tensile strengths than specified whereas 94½% of these had Brinell hardnesses below the values specified. Numerous reference tables are given. (17 References).

Hot-Bath Quenching at Low Temperatures of Eutectoid Carbon Steel. I. On the Structural Constituents After Quenching. I. Tatsukawa. (*Nippon Kinzoku*, 1957, **21**, Jan., 11–14). With a view to improving properties, 0·8%C steel was quenched in a hot bath below the M_s temp. The isothermal expansion curve shows an initial step corresponding with formation of athermal and isothermal martensite, and another associated with the lower-bainite reaction and stabilization of austenite. After quenching, the room-temp. constituents vary according to conditions, but structures of tempered martensite with lower bainite containing much retained austenite are possible.—K. E. J.

Hardening Liquids. (*Aciers Fins Spéc.*, 1958, Dec., 49–55). Aqueous and organic media and salt baths are briefly considered.

Features of the Process of Tempering after High-Frequency Hardening. I. N. Kidin, E. V. Astaf'eva and A. N. Marshalkii. (*Metall. Obra. Met.*, 1958, (9), 2–12). The non-uniformity of carbon distribution produced by rapid heating was examined by X-ray methods. Microstructures and microradiograms for a steel hardened at 1100° C with various rates of heating are shown, the non-uniformity increasing with the rate. Decomposition of the non-uniform martensite during tempering was also examined. Impact tests were carried out on steels after tempering at different temperatures for a range of times and the necessary rates of heating for optimum impact resistances are given and discussed.

The Influence of Tempering-Time on Some of the Mechanical and Physical Properties of Steel. J. L. Aston (*JISI*, 1959, **192**, Aug., 377-382), [This issue].

FORGING, STAMPING, DRAWING AND PRESSING

How To Get More from High Strength Steel. (*Steel*, 1958, **143**, Nov. 17, 128–129). Advice on producing high-strength wrought-alloy steel parts is given. This includes: (1) rolling and forging must be done without overheating or burning the metal internally; (2) machining must avoid tears and cuts, which produce stress-raisers; (3) the steel selected must maintain adequate ductility when heat-treated to high strength; (4) watch for embrittlement and cracking under plated finishes.—D. L. C. P.

Pointers for Purchasing Agents. (*Iron Age*, 1958, **182**, Aug. 21, 102–104). Operations which may be carried out in the forging shop to save machining costs, are considered. Examples are straightening, coining, planishing, punching, broaching, and ironing.

Static-dynamic Forging. P. Stanković. (*Met. Treatment*, 1958, **25**, Dec., 489–494). A machine built in the Machine Tool Dept. of Belgrade University is described and tests are reported. The theory of low speed (static) and high speed (dynamic) forging and of the combined process is given and the forces involved are illustrated and material flow, resistance to deformation and effects on structure are considered. It is shown that a balance between the press and forge methods gives good control of the flow of metal both in the direction of the force and in the perpendicular direction.

How Alloying Elements Affect Steel Forgeability. G. P. Contractor and W. A. Morgan. (*Iron Age*, 1958, **182**, Nov. 13, 142–146). A series of graphs are given which show the effect of steel composition on hot workability at various temperatures. The workability is measured by "the number of twists to failure" of a sample and this is considered a useful measure of forging properties. The effect of variations in the following in a selection of steels is shown: C, Mn, S, Si, Ni, Cr, Mo, N_2, Sn, Pb, P, O_2, Co, V, Ti.—D. L. C. P.

Forgeability of Steel Type 08kp with Additions of Titanium and Vanadium. N. D. Lomakin. (*Stal'*, *Supplt. Metallogr.* and *Heat Treatmt.*, 1958, 240–246). The best experimental results concerning yield after stamping (97·9%), showing the smallest number of displacement lines of least definition were given by sheets of steel type 08kp cast with additions of ferro-vanadium calculated for a vanadium content of 0·05% without reckoning the loss by oxidation.—L. H.

The Calculation of Forging Cooling Rates. D. P. Timo and R. M. Goldhoff. (*A.S.M.E. preprint*, 58-SA-4, 1958, pp. 11). A method is described taking into account the change of thermal properties and surface heat transfer with temperature and used for a 44 in. turbine forging and a 7 in. cylinder, with satisfactory results.

Metallurgical Examination of Forging-Quality Steels. J. D. Bunton. (*Met. Treatment*, 1958, **25**, Oct., 414–416). Synopsis of the NADFS Spring lecture. The assessment of the quality of steels for forging is discussed.

Rotary Forging Turns Profits Up. (*Steel*, 1958, **143**, Nov. 10, 122–123). The principles of rotary forging are described and examples of applications are given. In the process, the vertical work piece rotates while 4 horizontally opposed hammers beat the part to the desired contour. The shape may be made by having a shaped hammer-head, or controlling the length of the stroke, as the work passes the hammers. Hollow parts can be made by forging against a mandrel. Metal may be hot- or cold-worked Advantages are material saving, low tooling cost, speed, accuracy, considerable grain-flow and reduction of finish-machining —D. L. C. P.

Fabrication of Hollow Steel Propeller Blades. F. J. Altmann. (*Weld. Met. Fabr.*, 1958, **26**, Aug., 272–276). Precision forging, contour rolling, and the welding of the 16 structural components are described.

Materials Selector: Design Features of Metal Forms. (*Mat. Design Eng.*, 1958–59, *Reference Issue*, **48**, Mid-Oct., 294–323). Tables are given of the general properties of castings, forged, extruded and otherwise formed parts in iron and non-ferrous metals and plastics. These are then treated in greater detail and various tolerances and permissible cap depths, stretch data etc. are given. Spinning is also included and the properties of mechanical tubing.

Three New Ways to Make Precision Forgings, Extrusions. J. F. Murphy. (*Mat. Design Eng.*, 1958, **48**, Nov., 105–107). Brief descriptions are given of rotary forging, precision gear forging in screw presses and cross extrusion. Parts are illustrated and diagrams explaining the processes are given.

Improved Heat Treating Practice for Cold Roll Sleeves Made of Steel 9KhF. I. A. Brainin, A. I. Kondrashov, V. A. Kharchenko. (*Metallov. Obra. Met.*, 1958, (7), 43–46). The steel has 0·9%C, 0·65%Cr and 0·22%V from a basic O.H. furnace. Properties and particularly the best conditions for forging the metal were investigated. Appreciable grain growth and overheating start above 950° C. Increase of austenitizing temperature to 960° improves retained austenite stability, increases its minimum stability temperature in the pearlite range and lowers the martensite point. Forging in two upsetting steps gives finer grain and improves the structure. A 2-stage isothermal treatment cycle eliminates flaking without normalizing and gives a hard, but easily machined structure. The quenching and tempering practice was also worked out.

Plant Layout and Mechanization in the Press Shops. L. Walter. (*Sheet Metal Ind.*, 1958, **35**, Oct., 747–753). A method of planning lay-out on a plastic grid film, choice of power sources, press shot mechanization in an automobile body factory, conveyors and their control are discussed in an illustrated article.

Automatic Forging Press. (*Met. Treatment*, 1958, **25**, Oct., 424). A brief description of a 2500-t press developed by the Erie Foundry Co., Pennsylvania and producing 20 forgings per min. It is capable of 40, but this gives excessive die wear.

New Press Sweeps Hot Forging Into the Automation Age. (*Iron Age*, 1958, **182**, Oct. 2, 82–83). A fully automated four-station hot-forging press designed by the Erie Foundry Co., Pa., is described. With only one 'observer' it can produce up to 2400 finished forgings/h.—A. G.

Automation Triples Forging Output. (*Steel*, 1958, **143**, Sept. 29, 90–91). A 2500-t automated forging press developed by the Erie Foundry Company is described. Designed for high volume forging the press is said to triple output and requires only one operator.

Hot-Sizing Titanium and Heat-resistant Steel Parts. (*Machinery*, 1958, **93**, Oct. 1,

762–764). A brief account of presses developed in the U.S. is given.

Designing Metal Stampings. J. L. Everhart. (*Mat. Design Eng.*, 1958, **48,** Sept., 109–124). Design Manual No. 152. Austenitic stainless steels are included.

A Practical Manual for Stamping Design. Part II. Formed Stampings. C. C. Higgins. (*Mach. Design*, 1958, **30,** Aug. 7, 111–117). Deep-drawing is included in a general discussion of shapes, sizes, tolerances and finishes. Formed and bent parts are considered here.

Clogged Die Openings. F. Strasser. (*Iron Steel*, 1958, **31,** Oct., 497–498). Factors leading to the clogging of the slugs in the piercer openings of die plates, as found in small-hole piercing practice, are enumerated, and the mechanism of the process is described.

Improved Methods for Deep Drawing. (*Metal Progress*, 1958, **74,** Aug., 84–86). Points made at a conference on deep drawing involved advances in theory, materials and production methods.—D. L. C. F.

Preventing Coarse-Grained Fractures in Silicon-Chromium Steels. M. I. Vinograd and V. A. Zubrilova. (*Stal' Supplement Metallography and Heat Treatment*, 1958, 31–38). The steels used were of composition Kh9S2, 0·4%C, 2·43%Si, 8·66%Cr and Kh10S2M, 0·44%C, 2·41%Si, 9·3%Cr, 0·8% Mo. Coarse and non-uniform grain size is avoided by using a high final rolling temperature (900° C) and by making use of small reductions in drawing, up to 9% for Kh9S2 and up to 12% for Kh10S2M.—L. H.

ROLLING-MILL PRACTICE

Faster Roll Grinding Aided by Automatic Electronic Feed. (*Metalw. Prod.*, 1958, **102,** Dec. 26, 2265–2269). A new roll grinder for strip mills is described. Its capacity is 192 in between centres and on test it ground 0·024 in. from a 20½ in. roll of hardened forged steel 80 in. long in 15 min, to a 1½–2 micro-in. finish.

The Changing of Entire Stands for Roll Changing in a Rail Mill. V. V. Shakun. (*Metallurg*, 1958, (9), 26–27). The author describes the practice of changing entire stands in a 32 in. and 36 in. rail mill. Roll changing time was reduced from 3·5 to 1·8 h.

New Method for Increasing the Service Life of Roller Bearings and Reducing Specific Metal Consumption. A. M. Kovalevskii. (*Vestnik Mashin.*, 1958, (8), 3–9). Experimental investigations have established the following relation between the numerical value of the load at the point of contact and the service life of the bearings: $h = M/\sigma^S$, where h is the number of cycles up to the onset of fatigue damage (i.e. the service life), M, a constant for the material depending on various factors, σ, the load at point of contact. H + S the index of contact equal to 20/3 for a linear contact and 9 or 10 for a point contact. If, for example, the contact pressure could be halved, the service life would increase 100-fold. The method consists in altering the curvature of the surface of the bearings so that the load may be transmitted to the maximum number of rollers in the most efficient and uniform manner.—L. H.

Adjustment of the Free Play of the Shaft in Roller Bearings. B. G. Popovich and V. A. Roshet. (*Vestnik Mashin*, 1958, (8) 9–14). Methods of adjustment are suggested with tables for the numerical values of the free play of the shafts according to the various types of bearings and the load at points of contact.—L. H.

Rolling Mill Slippers. F. J. Gleve and R. J. Severson. (*Iron Steel Eng.*, 1958, **35,** May, 127–130). The authors emphasize the need for good design and the use of suitable materials to ensure good service of universal spindle bearing segments or slippers.

Operating and Mechanical Features of New 46 × 110-in., High-Lift Blooming Mill at Stelco. A. C. Gray and J. G. Mitchell. (*Iron Steel Eng.*, 1958, **35,** Aug., 123–131). This blooming mill, installed at the Hamilton works of the Steel Company of Canada, has been in operation since Jan. 1958, and is now rolling 300 ingot t/h. Constructional and operational details of the mill are given.

Operational Features of 45-in. Slab Mill and 80-in. Hot Strip Mill at Fairless Works. T. D. Clifford. (*Iron Steel Eng.*, 1958, **35,** Sept., 198–201). A description of the features of the plant which went into operation in 1953.

Pittsburgh Steel Completes 6-Stand, 30-in. Billet Mill. (*Iron Steel Eng.*, 1958, **35,** Oct., 145). A brief account of the installation of this mill in 20 days at the Pittsburgh Steel Co.'s plant.

First British Universal Beam Mill. New Lackenby Plant of Dorman Long (Steel) Ltd. (*Iron Coal Trades Rev.*, 1958, **177,** Oct. 3, 797–803). **Rolling Universal Beams.** (*Brit. Eng.*, 1958, **41,** Nov., 144–148). An illustrated account of the Lackenby rolling mill. **Universal Beam Mill at Lackenby.** (*Engineer*, 1958, **206,** Oct. 17, 605–608; Oct. 24, 637–640). With plans and illustrations. **British Mill Rolls Broad Flange Beams.** (*Engineering*, 1958, **186,** Sept. 26, 406–408). A description of Dorman Long's new Lackenby universal beam mill, with plan.

Bar Mill Design Reduces Need for Roll Changing. A. E. Lendl. (*Iron Steel Eng.*, 1958, **35,** Sept., 220–221). The layout and present capacity are described of a merchant mill installed in 1957 by Hoesch-Westfalenhütte A.G., Dortmund, which permits important reductions in or the elimination of downtimes due to roll and groove changing.

New Continuous Small Section Rolling Mill for a Large Rolling Schedule. N. Petrovits and H. H. Billen. (**Stahl Eisen*, 1958, **78,** Oct. 30, 1546–1566). The planning, design and construction of a small-section rolling mill of the Röchling'schen Eisen- und Stahlwerke at Völklingen is described. This involved shifting of the bed of the river Saar and of a main railway track as well as a main road. Two rolling trains are intended with a combined output of 30 000 t per month. The mill will eventually consist of 38 stands, three of them three-high. Auxiliary plant is included in the description.—T. G.

Republic Starts Operation on 11-in. Bar Mill at South Chicago. (*Iron Steel Eng.*, 1958, **35,** Aug., 142). A brief description of the layout of this mill is given. **Midwest Bar Mill, Now in Operation, Has Novel Features.** (*Blast Fur. Steel Plant*, 1958, **46,** Sept., 966–967). Republic Steel Corporation recently put a new 11-in. bar mill into operation at its South Chicago plant. The bar is not twisted or turned during reduction, and looping at the finishing stands is vertical.

Bar Mill Operates Automatically. C. A. Billante and G. Lambert. (*Iron Steel Eng.*, 1958, **35,** Sept., 203). A schematic diagram shows the operation of a 4-stand bar mill under fully automatic control at the Connors Steel Division of H. K. Porter Co. in Birmingham, Ala.

A Combination Mill for Rolling Rod, Bars and Strip. E. E. Davis. (*Iron Steel Eng.*, 1958, **35,** April, 101–105). The article describes a small flexible mill capable of rolling high quality products to closer tolerance. The mill comprises a 20-in. 2-high reversing bloomer, an 18-in., 3-high stand with tilting table, 8 stands of 14 in., 4 stands of 12 in. and 10 stands of 10 in. The mill can roll strip down from 14 in. to 3 in. and a variety of rounds down to about ¼ in.—M. D. J. B.

Automation of the First Rail Mill in the U.S.S.R. S. V. Gubert. (*Metallurg*, 1958, **7,** 29). The article reviews briefly the automatic control of the milling process and of the finishing train at Nizhne Tagil, (Urals).

The Shape of the Grooves and their Disposition in Existing Rail Rolling Mills. M. Musiał. (*Hutnik*, 1958, **25,** (7–8), 291–295). [In Polish]. The article discusses various methods of rolling rails used in U.S.A. and in Poland. In spite of the increased production of rails in Poland some methods used are 50 years old. The author proposes his own roll pass diagram for rails in seven stages; he regrets the absence of statistical data on the wear of rails, which should be supplied by the Polish State Railways.—S. S.

The Advantages of Beams in the New IPE Series. C. Perret. (*Acier-Stahl-Steel*, 1958, **23,** Oct., 420–422). An account of the new International series of lightweight beams with their structural properties.

Full Use of the Barrel of the Upper Roll in a 32 in. Stand. I. T. Lyzhenko. (*Metallurg*, 1958, (9), 28). Only half of the upper roll was formerly used. Grooves were cut in the unused half of the top roll and the position of the grooves in the centre and lower rolls was slightly adjusted by altering the dimensions of the collars. Now, only the lower and middle rolls are changed while the top roll is merely turned round, the other end being then used.—R. S.

Rod Mill Developments at Magnitogorsk (Urals). A. A. Davydov and N. V. Litovchenko. (*Metallurg*, 1958, **7,** 30–31). An account is given of the improving of a continuous rod mill imported for Magnitogorsk in 1938 and the present pass design is shown for rolling 0·255, 0·275 and 0·315 in. rod.

Contour Rolling Complex Shapes. F. J. Altmann and R. D. Libert. (*Metalw. Prod.*, 1958, **102,** Dec. 12, 2184–2186). Typical aircraft and jet turbine parts are shown.

New Rod Mill in Germany Embodies Many Novel Features. (*Wire Wire Prod.*, 1958, **33,** June, 644–647, 705–706). The mill here described is that built by Schloemann A.G., installed at the Kloeckner Hüttenwerk Haspe A.G., and put into operation in 1955–56. Included are full plant data and numerous illustrations. Notable features are, individual strand rolling in finishing trains in which vertical and horizontal stands alternate and are fitted with automatic loop control for tension-free rolling; mill motor regulation restricting dynamic speed difference during sudden surges to within 1·5%, with a 0·15 sec. recovery time; sequence control of mill motor speeds with speed setting to within 0·1%.—J. G. W.

Merchant and Wire Rod Mill. Fully Continuous Plant at Klockner. L. Walter. (*Iron Steel*, 1958, **31,** Oct., 485–490). The fully-continuous mill of Klöckner-Hüttenwerk Haspe A.G., of which the merchant mill went into operation in 1955, and the wire rod mill in 1956 is described in detail.

Modern Wire Rolling Mills. O. Wilmes. (*Wire*, 1958, Aug., 15–18). The ARBED strip and rod mill built by Demag is described and illustrated.—J. G. W.

Planning the Design, Layout and Construction of a Heavy-Plate Rolling Mill. W. Hilterhaus. (**Stahl Eisen*, 1958, **78,** Oct. 30, 1525–1536). The work involved in the planning of a heavy-plate rolling mill consisting of a four-high stand and a vertical stand for slabs of 2–40-t at the Henrichshütte of Ruhrstahl A.G. at Hattingen is described. Design of the stands, in particular the rolls and their diameter, is discussed in detail. Layout of the plant and of the soaking and cooling pits are shown in diagrams. Models of the plant were prepared, these were later used for the training of operators.—T. G.

Appraisal of Design and Type of Operation of a Heavy-Plate Rolling Mill on the Basis of Operational Results. P. Diestel. (**Stahl Eisen*, 1958, **78,** Oct. 30, 1536–1546). The mill, described in the preceding abstract, has been in operation for 20 months, during this time the throughput was 30 000 t per month. Operation, output, costs for working and consumption of energy and water, life of the rolls, control equipment including TV apparatus and quality control are dealt with.

Design and Construction of a New Heavy-Plate Mill. H. Matthaei. (**Stahl Eisen*, 1958, **78,** Oct. 30, 1609–1611). The modernization of the old 2500-mm three-high heavy-plate mill at Hüttenwerk Oberhausen into a 3300-mm four-high reversing mill is described. The mill produces plates 5–20 mm thick and up to 30 m long, the output is 30 000 t per month. Slabs up to 7·2 t are rolled. The vertical upsetting stand is particularly powerful having rolls of 1 m diameter and exercising a roll pressure of 300 t. Thickness measurement of the plate is carried out with X-ray equipment.—T. G.

Guiding Principles for Rolling High-Alloy Steels. H. Müller. (**Stahl Eisen*, 1958, **78,** Oct. 30, 1611–1613). An illustrated review

of the general layout and pass sequence of the new rolling mill for rolling section of high-alloy steels of the Società Nazionale Cogne at Aosta, Italy. The peculiarities of rolling high-alloy steels are discussed.—T. G.

The Hot-Rolling of Mild Steel with Large Reductions. I.—Spread and Elongation in Hot Rolling. Z. Wusatowski. (*Berg.-Hütten. Monatsh.*, 1958, **103**, July, 136–142). The relationships between the degree of spread, elongation and upset and the possible reduction in the hot-rolling of mild steel are investigated theoretically and in the light of results from an experimental 2-high mill. **II.—Forward Slip and Rolling Speed.** (Aug., 153–160). The amount of forward slip is influenced by the reduction, coefficient of friction, the shape of the test piece and the temperature. The influence of these factors is assessed. **III.—The Flow of Metal and Structural Appearances in Hot Rolling.** Z. Wusatowski and B. Hoderny. (Sept., 175–180). A series of experiments on the rolling of mild steel, in relation to the flow of metal, showed that the use of heavy reductions in hot rolling had no marked harmful effect on the rolled material.

Some Practical Solutions to Hot Mill Problems. R. A. Smith and A. D. Patton. (*Iron Steel Eng.*, 1958, **35**, July, 92–96). Five problems connected with the design and operation of a 96-in. continuous hot-strip mill are discussed. They concern the modification of a reheating furnace to take larger slabs, reduction of furnace shutdown periods, and various mill modifications.

Semi-Continuous Hot-Strip Rolling with Automatic Programme Control. A. Hegarty. (*Sheet Metal Ind.*, 1958, **35**, Aug., 623–625). A 44-in. semi-continuous hot-strip rolling mill with a fully programmed control system on the reversing rougher was taken into production by Jones and Laughlin Steel Corp., U.S.A. All an operator has to do at the beginning of a rolling schedule is to bring the slab coming from the primary mill on to the entry table and to place an IBM punched card containing information on the requirements for that slab into a reader unit. The card will supply nine types of information. The mill operation including six four-high finishing stands is briefly described.—T. G.

High Production in Flat Rolling. C. S. Ball and C. W. Starling. (*Sheet Metal Ind.*, 1958, **35**, Oct., 767–774) A continuous strip mill is considered. Location is briefly referred to and flow through the mill is analysed in terms of material transport and buffer stocks, and breakdowns and their effects are discussed. Trends towards simplification, especially in the direction of direct steel production and the manufacture of precoated sheet are finally evaluated.

Recent Hot Strip Mill Roughing Trains. J. H. Greiner. (*Iron Steel Eng.*, 1958, **35**, Sept., 151–159). Recent trends in roughing stand drives on hot strip mills include the use of 6 000-h.p. synchronous motors in roughers following the broadside stand, and 3 000–5 000 h.p. wound-rotor motors on the latter. The use of inching and dynamic braking equipment in reducing down times is discussed.

Modifications Double Capacity of Empire-Reeves Hot Strip Mill Line. (*Iron Steel Eng.*, 1958, **35**, Oct., 161–162). By the addition of an intermediate roughing stand, a flying shear and a down coiler the rolling cycle was modified so as to double the capacity of the line.

Selection of Electrical Equipment for Temper and Skin Pass Mills. J. E. Peebles, J. J. Remley and R. B. Eggleston. (*Iron Steel Eng.*, 1958, **35**, Sept., 115–128). The advantages of 1 and 2-stand temper mills are compared. The higher cost of the 2-stand mill is only justified when high output is required. Data for selecting drives and controls are given.

Temper Pass Mill Provides Large Power and High Speed. A. F. Kenyon. (*Iron Steel Eng.*, 1958, **35**, Oct., 156). Data are given of the drive units employed on this installation.

Planned Maintenance Techniques for a Hot Strip Mill. J. R. Kennedy. (*Iron Steel Eng.*, 1958, **35**, Oct., 139–144). Factors, mainly suitable inspection and recording, which reduce maintenance delays at a hot strip mill are described, with reference to experience at the Pittsburgh Works Division of Jones and Laughlin Steel Corp.

How to Design Conveyor Rolls for Drive or Drag. L. H. Austin. (*Prod. Eng.*, 1958, **29**, Sept. 15, 78–79). Arrangements for moving metal strip are shown.

Methods of Control of Strip Dimensions on Hot Strip Finishing Mills. R. A. Phillips. (*Iron Steel Eng.*, 1958, **35**, May, 100–106). This paper discusses the methods for closer control of strip dimensions in tandem hot strip finishing trains as a means of improving the products of the cold rolling mills.

Operations Research in a Metal Rolling Mill. A. P. M. Purdon. (*Met. Treatment*, 1958, **25**, Dec., 503–507). The descriptions refer to the rolling of Cu, but are sufficiently general to be of interest in steel mills as indicating the use of the operational research method. The analysis of the process from all directions bearing on efficiency and economics and the kind of conclusions to be expected are indicated.

Electronics Regulates Strip Mill. (*Steel*, 1958, **143**, Oct. 27, 94). A brief account of the problem (non-uniform heat retention) at the Geneva (Utah) works of U.S. Steel and the gauges and other controls installed for its solution is given.

WELDING AND FLAME CUTTING

What do the Russians Weld With? P. Trippe. (*Prod. Eng.*, 1959, **30**, Jan. 19, 64–67). D. Swan. (65). D. Vvedensky. (66). R. J. Gasser (67). The electroslag, submerged-arc and resistance welding processes, especially by automatic machines, are evaluated with comments by the other authors.

Generalities and Evolution in the Economic Domain of Electric Arc Welding. W. Bonhomme and G. Jasmin. (*Rev. Univ. Min.*, 1958, **14**, Oct., 372–377). Improvements in welding electrodes and their effects on the economics of electric arc welding are discussed.

Submerged Arc Welding. L. Adam. (*Rev. Univers. Mines*, 1958, **14**, Oct., 317–328). An account of the theoretical basis of the process, a description of the development and operation of the submerged arc welding process.

Submerged Arc Welding Helps Steel Plant Maintenance. J. Angus. (*Iron Steel Eng.*, 1958, **35**, Nov., 132–137). Several examples of steel plant welding are given which illustrate the flexibility of submerged arc welding for plant maintenance.

Electric Arc Welding under a Protective Atmosphere. A. Moreau. (*Rev. Univers. Mines*, **14**, Oct., 329–342). Some modern applications of welding under protective atmospheres are described, with ferrous and non-ferrous examples of the use of A, reference being also made to the use of CO_2.

Automatic Welding under a Protective Atmosphere of Carbon Dioxide. F. G. Danhier. (*Rev. Univers. Mines*, 1958, **14**, Oct., 343–349). A description of automatic and semi-automatic welding under CO_2, with particular reference to the Arcos arc welding process.

Industrial Applications of Magnetic-Flux Gas-Shielded Arc-Welding. R. T. Telford and F. T. Stanchus. (*Weld. J.*, 1958, **37**, Aug., 771–778). The authors describe a number of different production applications of magnetic-flux welding and summarize economic as well as technical advantages of the process. The process itself was described earlier.—T. G.

The Use of Protective Gas as Protection Against Oxidation in the Welding of Alloy Steels. J. Wingerath and K. Boeckhaus. (*Schweissen Schneiden*, 1958, **10**, Dec., 471–475). The causes and effects of oxidation at the weld root are examined, methods of preventing oxidation discussed, and the use of "generator gas" (a mixture of N_2 and H_2) as a protective gas for welding described.

The Welding of Screwed Couplings on Cast Iron Pipelines with Screwed Fittings. M. Jürgens. (*Schweissen Schneiden*, 1958, **10**, Oct., 411–412). Practical details are given of the repair by welding of cracked threaded fittings for cast iron pipelines.

Diffusion Welding *in Vacuo*. N. F. Kazakov. (*Avtom. Svarka*, 1958, (8), 6–18). The effects of pressure, temperature, surface cleanness, composition and properties of metals and alloys, on their weldability *in vacuo* are examined. A special apparatus is described for such welding at high temperatures (up to 1300° C). The experiments showed that the strength of welds in metal-base metallo-ceramic alloys is dependent on the Co content of the alloy and grain size, increasing in proportion to the fineness of the grain. An increase in pressure leads at low temperatures to an inappreciable rise in strength, but at high temperatures to a rapid rise, but pressures > 2·5 kg/mm² are not beneficial. If components are heated in a non-oxidizing atmosphere, not only oxidization, but also the formation of brittle layers and other defects in the weld zone is avoided. The main advantage of the method is that a joint is formed without any noticeable change in the properties of different metals which are welded.—M. T.

Stability of Electroslag Welding in Relation to the Properties of the Flux. Yu N. Gotal'., skii. (*Svarochnoe Proizvodstvo*, 1957, (3) 1–3). The character of the slag required for the process is discussed such as low interaction with the metal, conductivity and arc suppression. Experiments with various fluxes are reported concerning especially the effects of fluorite.

Lightening and Welding: Mechanically Welded Steel Parts. A. Ogus. (*Rev. Gén. Méc.*, 1958, March, 175–178). The author discusses some practical points in welding and the forces in the weld area. He then goes on to consider the economic aspects of welding.—R. P.

Positioners and Manipulators. S. G. Maine. (*Eng. Foundryman*, 1958, **24**, Aug., 48–52). General recommendations for welding practice with a view to increase in output.—T. G.

Cold and Hot Electrodes and Their Use. M. Werion. (*Prat. Soud.*, 1958, **12**, Apr., 67–70). A description is given of the properties and uses of various electrode coatings.

Welding with OK B-Electrodes. M. Puschner. (*E.S.A.B. Svetsaren*, 1958, **23**, (2), 25–38). Welding of dissimilar metals and carrying out repairs with OK B-electrodes is described.

Present-day State of Development of Coated Electrodes for Manual Arc Welding of Ferrous Materials. K. L. Zeyen. (*Draht*, 1958, **9**, June, 211–215; Aug., 311–316). German, American, and British Standard Specifications are summarized and compared, and mention is made of high-performance electrodes and other recent developments. In addition to technical information, production statistics of the various types of electrodes are quoted and commented on.—J. G. W.

The Weldability of Low Alloy Manganese Steel 14-G-2. A. Ya. Brodskii. (*Avtom. Svarka*, 1958, (10), 44–54). Electrodes for welding manganese steel are examined and changes taking place in the steel during welding and the susceptibility of the steel to stress concentrations and stress peaks were investigated. No cold shortness was observed down to — 70° C. The properties of both the steel and the welded joints were good, provided the steel was satisfactorily deoxidized during its production.—M. T.

Sound Welds Reported for Resistance-Welded Low Carbon Steel Pipe in Evaluation Tests. R. C. Bertossa. (*Western Metalw.*, 1958, Oct., 48–50). The equipment and process for the high-frequency resistance welding and induction normalizing of low-carbon steel pipe are described. The pipe was subjected to API tests for line pipe and to the ASTM test for resistance-welded steel pipe. These tests included chemical analysis, hydrostatic tests, transverse tensile tests and metallurgical examination. Sound, ductile welds were produced by the method mentioned.

Special Structural Features of Spot Welds between Different Metals of Small and Unlike

Thicknesses. V. E. Moravskii and A. A. Rossoshinskii. (*Avtom. Svarka*, 1958, (10), 90–96). A metallographic analysis was made of such spot welds. It was found that fusion is not essential to produce good quality welds. If fusion does not occur, crystalline transformation or recrystallization takes place. During the welding of materials which have no tendency to form brittle hardened layers, in the area around the spot, fine-grain structures are formed which have high plastic and strength properties. Almost all combinations of metals and alloys may be joined by this method.—M. T.

Effect of Niobium on the Structure of Welds in Austenitic 18-8 Steel. G. G. Mukhin and N. Yu. Pal'chuk. (*Avtom. Svarka*, 1957, **10**, (2), 1–10). Up to 1%Nb is often added to prevent intergranular corrosion. In solid solution it promotes ferrite formation and a second phase of combined Nb is also observed. Various carbides have been identified in higher Nb steels. In welds it is predominantly a carbide forming element and high contents may cause hot cracking. The work described was on sigma-phase formation and various ageing treatments were used. Metallographic and X-ray studies were made and the phases appearing were identified. With $> 2\%$Nb, Fe_2Nb_2 and $Fe_4Nb_2Si_3$ were formed during welding in addition to carbide and H-phase in heat treatment. The H-phase reduces ductility, it is isomorphous with χ-phase found in Mo steels. Sigma-phase forms on prolonged tempering at 650° C with $< 2\%$Nb and H-phase with $> 2\%$, both lowering impact strength and ductility.

New Welding Processes. C. W. J. Vernon. (*Weld. Met. Fabr.*, 1958, **26**, Sept., 328–337). An illustrated review is given of CO_2 processes, dual-gas shielding, ultrasonics, bonded fluxes, electron-beam fusion, automatic pipe joining and H.F. resistance welding methods.

Welding at the Service of Lacq Gas-Field Working. P. Berthier, R. Enous, J. Souviron, R. Terminet and G. Reynes. (*Soud. Techn. Connexes*, 1958, **12**, Sept.-Oct., 357–364). In the recently discovered gas fields at Lacq, problems have arisen due to the high pressure (670 kg/cm²), temperature (140° C), high H_2S content (up to 15%) and high output of the gas. Welded parts are inspected with special care and all assemblies are stress-relieved. The gas washing towers are made of a 0·3%C steel of good corrosion resistance under prevailing conditions. After stabilizing, the lower sections are fitted with a 17Cr–12Ni–2/2·5 Mo steel lining 3 mm thick. The recovery plant carrying purified gas does not present special difficulties.—R. P.

CO_2 Welds Thin Steel Parts. (*Metalw. Prod.*, 1958, **102**, Nov. 7, 1961). Welding of 22-gauge mild steel sheet for automobile construction with a shielded arc is briefly described, making use of the new "Dynamic Reactor" power supply.

CO_2 Welder Makes Casting Repair Easier. J. J. Chyle. (*Steel*, 1958, **143**, Nov. 17, 110–112). The applications and advantages of CO_2 welding are considered. Its chief use in the foundry is for carbon grades, although it can be used on some low-alloy steels. Advantages: low cost (largely due to cheap gas), visible arc, no flux, deep penetration, high quality (e.g. elimination of hydrogen embrittlement) and suitability for automatic operation.—D. L. C. P.

Fabricator Sets Up "Countdown" To Ensure Top Quality Welds. R. F. Gurnea. (*Iron Age*, 1958, **182**, Nov. 27, 65–68). The production of 16 in. diameter stainless steel elbows for a nuclear reactor cooling system is described. The main emphasis is on the extensive testing that was carried out after each production step, to ensure a perfect product.—D. L. C. P.

Salvage of Iron Castings by Welding. J. C. Pellegrino. (*Foundry Trade J.*, 1958, **105**, Nov. 6, 573–574). Arc welding shows economies over gas welding. Work done by General Motors is described on cylinder-head castings cut into 6 parts and holes and cavities bored into the surfaces. Arc and gas welding was compared by methods described. Time taken and labour costs were reduced and the greater cost of arc-welding equipment was largely offset by decreased maintenance.

Post-Weld Treatment of Welded Units for the Relief of Stress. S. J. Watson. (*Weld. Met. Fabr.*, 1958, **26**, Sept., 318–322). Heat treatments are described, either of the whole part or localized, and tests are outlined for residual stresses.

The Improvement of Welded Structures by Subsequent Forging Operations. W. R. Wollering and E. H. Lundby. (*S.A.E. Trans.*, 1958, **66**, 261–266). SAE 4340 is usually lower in ductility at welds made from electrodes of comparable hardness. Hot-forging is shown greatly to overcome this difficulty. The process and the structures produced are described. Economies are claimed by this method of fabrication.

Magnetic-Force Welds Join Vinyl-Clad Steel. F. R. Park. (*Metalw. Prod.*, 1958, **102**, Nov. 28, 2079–2082; *Prod. Eng.*, 1958, **29**, Sept. 15, 82–85). A process using an air cylinder for initial holding and an electromagnet for controllable precisely-timed welding force in conjunction with a welding current lasting only $\sim 1{\cdot}5$ millisec. to protect the coating is described.

Magnetic-Flux Gas Shielded Arc Welding. J. E. Dato. (*Iron Steel Eng.*, 1958, **35**, Sept., 160–165). This process is intended to fill the gap between manual covered electrode welding and fully automatic production units. The principle, operation, and some applications of the process are described.

Fabricating Plastic-Clad Sheet Metals. (*Weld. Met. Fabr.*, 1958, **26**, Aug., 294–297). Welding methods are described.

Co-operative Investigation of a New Welding Electrode for Stainless Steel. R. D. Wylie. (*A.S.M.E. preprint*, 58-Met-6, 1958, pp. 8). Trials with Croloy 16-8-2 and the properties of the welds and weld metal are reviewed.

A Comparison of Methods for Assessing Spot Weld Ductility in Sheet Steels. E. J. French and A. A. Wells. (*Soud. Techn. Connexes*, 1958, **12**, Sept.-Oct., 369–392). Tensile, shear, torsion and hardness tests were carried out on spot welds deposited under various conditions on four steels containing 0·04 to 0·20%C, the thickness ranging between 1·2 and 2 mm. The results gave relations between the various tests and these results were compared with theoretical values. These relations can be used as a new basis for assessing spot weld ductility. (17 references).—R. P.

Metallurgical Function Stability of Arc Welded Joints in Unalloyed and Low Alloy Structural Steels. T. M. Norén. (*Jernkont. Ann.*, 1958, **142**, (11), 661–719). Weldability and functional stability are discussed as concepts the latter covering the whole construction. Factors bearing on these, particularly brittle fracture, are discussed and crack initiation as well as embrittlement not causing cracking are considered. The concept of nominal cleavage strength is then introduced and types of stability are distinguished. (36 refs.).

A New Method for the Detection of Internal Welding Stresses. W. Rädeker. (*Schweissen Schneiden*, 1958, **10**, Sept., 351–358). Internal stresses due to welding can be detected with great accuracy by digesting specimens in 60%Ca–NH_4 nitrate solution. A form of sample is described, by the aid of which it is possible to examine the effect of welding conditions. Several examples of the applicability of the method are given.

Co-operative Research Regarding the Reproducibility of the Controlled Thermal Severity Weldability Test for Low-Alloy Steels. I.I.W.-Commission IX. (*Soud. Techn. Connexes*, 1958, **12**, Sept.-Oct., 345–353). The test described is a restrained fillet weld test developed from earlier methods. The fillet weld is deposited under predetermined conditions, the transverse section of the weld being examined for cracks in the heat-affected zone of the parent plate. Results supplied by five laboratories show that the test is satisfactory for scientific investigations of the phenomena associated with heat-affected zone cracking.—R. P.

Prevent Hot Cracking in Welding. (*Metalw. Prod.*, 1958, **102**, Oct. 31, 1908). The usual causes and protection against them are briefly given.

Weld Flaw Evaluation. S. T. Carpenter and R. F. Linsenmeyer. (*Weld. Res. Council, Bull. Series* 42, 1958, Sept., pp. 37). A final report is presented of studies on the evaluation of the ability of weld flaws to initiate brittle fracture. The theory of embrittlement is reviewed and static tests on flawed butt welds, static and dynamic tests on small flawed butt welds with and without residual stress and static tests on weld flaws with a controlled high residual stress field are reported. Effects of low temperature and other factors in embrittlement are considered and crack formation is discusssed. The importance of residual stress is pointed out.

Joining of Metals in Solid State by Ultrasonic Vibrations. Yu. I. Kitaigorodskii, M. G Kogan, V. A. Kuznetsova, N. N. Rykalin and L. L. Silin. (*Izvest. Akad. Nauk. Otdel. Tekn.*, 1958, (8), 88–90). Non-ferrous examples are given.

Wade Furnaces for Brazing Stainless Steel and Heat-treating Jet Pipe Assemblies. (*Machinery*, 1958, **93**, Sept. 17, 645–647). Equipment installed by J. Lucas Ltd. at Burnley is described.

The Hot Strength of Brazed Joints Made with Various Filler Metals. B. Blanchet. (*Soud. Techn. Connexes*, 1958, **12**, March-April, 101–106). The author discusses tests and results of investigations of the hot strength of joints brazed with five different fillers. Shear strengths in presence and absence of creep are given. Temperature limits in relation to mechanical behaviour are proposed.—R. P.

Investigation into the Weldability of Flame Treated Surfaces. H. Schmidt. (*Schweissen Schneiden*, 1958, **10**, May, 167–173). The paper discusses the principles and procedure used for testing the weldability of untreated and flame-treated sheet. By means of mechanical, metallographic and non-destructive tests it was shown that the weldability of certain sheets is not impaired by the treatment. The use of basic electrodes causes slag inclusions and further tests in this direction are advocated.—R. P.

Approach to Equilibrium Conditions in Molten Metal and Slag during Arc-Welding: Alumina-Silica Fluxes. W. M. Conn. (*Schweissen Schneiden*, 1958, 5, May, 174–180). The method of producing fluxes for submerged arc welding and the conditions for testing are discussed. Results of tests are given and the rate of travel of a gas bubble through molten metal and slag are considered. This is followed by considerations of the influence of viscosity, surface tension, density, rates of crystallization and melting, thermal conductivity and also diffusion. (19 references).

Evaluation of Spot Welds Made Through Primers and Sealers. B. A. Schevo. (*Weld. J.*, 1958, **37**, Aug., 799–802). A number of very different steels were primed and sealed and the effect tested by corrosion behaviour. Spot-welding was performed satisfactorily through one primer and four sealers. Sealers satisfactorily prevent water leaks on joints which had sealers applied to the facing surfaces before spot welding.—T. G.

Radiographic Control of Welded Structures. A. R. Rubio. (*Cien. Tecn. Solda.*, 1958, **8**, Sept.-Oct., 1 (115)). The principles of radiographic non-destructive testing are outlined, and an example of the examination of a welded joint is discussed.

Comparison of Different Methods of Evaluating the Ductility of Resistance Spot Welds. E. J. French and A. A. Wells. (*Rivista Ital. Soldatura*, 1958, **10**, Sept.-Oct., 191–208). [In Italian]. The report describes experiments to compare the merits of various test methods for spot welds. The authors claim that tension tests show a better degree of ductility discrimination than shear and torsion tests. The ratio of experimental to theoretical tensile strength read in conjunction with certain other tests gives a good indication of spot weld ductility. Considerable tech-

nical data are given in table form. (17 references].—M. D. J. B.

Equipment for Torsion Testing Resistance Spot Welds. T. M. Roberts. (*Rivista Ital. Soldatura*, 1958, **10**, Sept.-Oct., 188–190). [In Italian]. The author describes a new type of torsion testing machine for spot welds. The machine is power-operated and records graphically torque and angle of failure. The design, operation and calibration of the machine are discussed and records of typical tests on four different steels are given.

The Connection Between the Hardness and the Microstructure of the Fusion Zone of Austenitic Joints in Pearlitic Steels and the Chemical Composition of the Steel and the Joints. L. S. Livshits and L. P. Bakhrakh. (*Avtom. Svarka*, 1958, (10), 81–85). The hardness, and probably the brittleness also, of the fusion zone of pearlitic steel with an austenitic joint is decreased by reducing the carbon content of the steel and the Cr content of the austenitic joint, and on the other hand by increasing the content of Ni in the weld and of elements in the steel which form more stable carbides than those of Cr. In order to reduce the hardness and improve the homogeneity of the fusion zone of pearlitic steel with an austenitic joint, intended for operation at elevated temperatures, 18–8 austenitic electrodes should be used, not containing Ni, Ti, Mo, W, for the welding of steels alloyed with Mo, W or V, either separately or jointly, in such quantities as will ensure conversion of the carbon in the steel into the relative carbides.—M. T.

The Problem of the Primary Crystallisation of the Weld Metal during Fusion Welding. M. I. Razikov. (*Avtom. Svarka*, 1958, (8), 60–62). During the welding of steels the molten austenitic grains of the base metal along the line of fusion form bases for the growing columnar dendrites of the weld metal. Thus the grain size of the metal of welded joints is to a large extent dependent on the grain formation of the base metal along the line of fusion at the moment when the dendrites start to grow. To reduce the dimensions of the dendrites the grain growth of the base metal should be restricted to a minimum.—M. T.

Research into the Distribution of Hydrogen in Austenitic and Ferritic Welded Joints in Medium Alloy Steels. A. M. Makara, V. I. Lakomskii and I. P. Zhovnitskii. (*Avtom. Svarka*, 1958, **11**, (11), 16–31). The authors define ferritic and austenitic welds as those with a matrix of α- and γ-iron respectively. It was found that in ferritic joints in hardenable steels H_2 is primarily dispersed from the weld metal, and in austenitic joints from the heat-affected zone. H_2 distribution is almost uniform along the length of a weld. Over a cross-section in austenitic welds in areas adjacent to the heat-affected zone the H_2 content rises sharply, but in ferritic welds it falls slightly, and in both types is somewhat higher in the surface layers than in the heart of the weld. In areas of the heat-affected zone in contact with the weld metal in both types the H_2 content is many times greater than that of the base metal and close to that of the weld metal with which it is in contact, but falls abruptly in proportion to the distance from the fusion zone boundary. Under identical conditions the H_2 content in the heat-affected zone of austenitic welds is about six times as great as in ferritic welds. These discrepancies are explained by greater H_2 solubility in the molten metal of, and the high solidification rate of the peripheral sections of, austenitic welds, and the rate and duration of H_2 migration in ferritic welds from the fusion zone boundary.—M. T.

Cracks in Alloy Steel Welds in the Zones Affected by Temperature Changes. H. Sekiguchi and T. Kobayashi. (*Riv. Ital. Soldatura*, 1958, **10**, May-June, 108–114). The authors distinguish four types of cracks: Toe cracks, underhead cracks, transverse and longitudinal cracks—on transverse and longitudinal sections of small square specimens of Ni–Cr–Mo steel on which weld beads had been deposited. Based on a large number of cracks investigated, the authors claim that hydrogen has a great influence only on underhead cracks. Times between the end of welding and the beginning of crack-formation have been recorded and it is shown that these lines can be related to martensite content in the joint. Finally, three welding procedures using heat treatment to prevent these cracks are described.—M. D. J. B.

Cracking Due to Welding in the Transformation Line of Air-Hardening Alloy Steels. H. Sekiguchi and T. Kobayashi. (*Cien. Tecn. Solda*, 1958, **8**, Sept.-Oct., 2 (24). Four types of fissure were distinguished in welded Ni–Cr–Mo specimens; their origin is discussed, and methods of prevention suggested.

PROTECTIVE COATINGS

Hot Galvanising Practice in U.S.A. A. T. Baldwin and W. H. McMullen. (*Electropl. Met. Fin.*, 1958, **11**, Aug., 274–278; Sept., 319–322). The more important aspects of American practice which have led to production of better coatings, and reduction in the amount of zinc lost as dross are considered. Wire galvanizing, sheet and strip galvanizing and the nature of the zinc coating are briefly described.

Some Considerations in Relation to the Hot-Dip Galvanizing of Guard Rails. H. Bablik, F. Götzl and E. Nell. (*Met. Fin. J.*, 1958, **4**, Aug., 297–300). Zinc alloy formation in the thick coating of pearlitic steel and the effects of previous filing, which greatly accelerates the reaction are discussed and illustrated as are the effects of Al in the bath.

RTB's Continuous Galvanising Line. (*Brit. Steel.*, 1958, **24**, Aug., 268–269, 271). An illustrated account of the train producing "Speltafast." **Automatic Galvanising Plant.** (*Engineering*, 1958, **186**, July 18, 78–79). An account of the Armco-Sendzimir installation at Ebbw Vale is given. **Continuous Galvanizing of Steel Strip.** (*Metallurgia*, 1958, **58**, Aug., 76–79). An account of the Ebbw Vale installation. **New Galvanizing Line: Further Details of R.T.B. Plant.** (*Iron Steel*, 1958, **31**, Aug., 390).

Strip Preheat Ends Galvanizing Pot Downtime. E. J. Udick and C. A. Turner, jun. (*Iron Steel Eng.*, 1958, **35**, Aug., 145–146, 149). By the use of a special direct-fired strip reheating furnace, without oxidation, strip fluxing is eliminated and the required Zn bath temperature is maintained without external pot firing.

New Radiant Tube Furnaces for Processing in Galvanizing and Patenting Lines. W. D. Bawden. (*Wire Wire Prod.*, 1958, **33**, Aug., 885–887, 922–925).

Electrodeposition of Iron–Zinc Alloys from the Pyrophosphate Bath. Vasanta Dree and T. L. Rama Char. (*J. Sci. Indust. Res.*, 1958, **17B**, Oct., 439–440). Bath compositions and operating conditions are given. The deposits are being further examined.

Preflux Solutions from Galvanizers' Sal Skimmings. P. M. Sullivan, D. H. Chambers and P. J. Berney. (*Met. Ind.*, 1958, **93**, Dec. 12, 493–496). Both water and acid leaching processes are effective. Control of pH, density and Cl content is sufficient for control.

Sprayed Electrolytic Zinc Coatings. R. Bourcelot. (*Met. Ind.*, 1958, **93**, Nov. 28, 453–454). The "Tropicalization" process is described. Supplementary coating and painting is also considered.

New Lining Methods. W. Füller. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 557–559). Linings of vessels of all kinds in the chemical industry must be resistant to chemical, thermal, and mechanical attack, and to swelling. The latter term is defined in detail. The use of plastic-impregnated graphite plates is advocated.—L. J. L.

Third Electrolytic Tinning Line Installed at Pittsburg, California. (*Blast Furn. Steel Plant*, 1958, **46**, Aug., 872–873).

On the Origin of "Coloured Spots" on Tinplates. N. I. Sandler and G. G. Solov'eva. (*Stal' Supplement Metallogr. and Heat Treatmt.*, 1958, 247–250). The bright surfaces of plates prepared for tinning sometimes show matt spots which remain uncovered by the tin. Investigations, with special instruments are described. The findings show that the spots appear after annealing the rolls with the steel sheet and suggest a thin oxide layer, difficult to remove by etching, but easily removable with polishing. Below this layer the metal structure is normal.—L. H.

Whats Inside the Coil. R. R. Davisson. (*Steel*, 1958, **143**, Sept. 8, 106–110). The growing tendency to purchase tin plate in coils has led to the development of automatic scanning equipment to check quality before coiling. A typical system records total footage with a light top or bottom coating, pin-holes and other defects, total footage of prime, menders and waste, coil number and weight, order number and date.—A. G.

The Influence of Surface Characteristics of Tinplate on De-wetting by Lacquer. S. C. Britton and K. Bright. (*Sheet Metal Ind.*, 1958, **35**, Sept., 679–684).

Automatic Hydraulically Operated Electroplating Plant. (*Met. Fin. J.*, 1958, **4**, Oct., 400–402). A new Cd-plating plant at Slade Green, Kent, is described.

High Speed Brass Plating. E. J. Roehl. (*Electropl. Met. Fin.*, 1958, **11**, Sept., 299–301). Production of brass-plated steel strip is described.

The Electrodeposition of Iron–Molybdenum Alloys. L. O. Case and A. Krohn. (*J. Electrochem. Soc.*, 1958, **105**, Sept., 512–520). A special cell rotating the cylindrical cathode between wiper blades, which greatly increases efficiency, was devised and conditions for producing bright adherent deposits are laid down.

Metal Spraying in Relation to the Corrosion Process. J. R. Franklin. (*Corros. Techn.*, 1958, **5**, Oct., 323–325). Spraying steelwork with another metal (normally Zn or Al) gives cathodic protection which because of the thickness of coating is long-lasting. Various applications and the type of equipment used are discussed.—A. G.

Materials Selector: Finishes and Coatings. (*Mat. Design Eng.*, 1958–59, *Reference Issue*, **48**, Mid-Oct., 259–277). Tabulated data on conversion coatings (anodic, chromate, phosphate and oxide) on iron and non-ferrous metals are given. Cr and Ni plating baths are given and pretreatments suggested. Zn and Cd baths, sprayed coatings, mechanical finishes for stainless steel (and Al and Cu), general properties of alkyd, alkyd-amine, alkyd-urea, styrene-alkyd, Alkyd-phenolic, alkyd-silicone and phenolic coatings; also silicones, epoxy and copolymer epoxy, vinyl, vinyl-alkyd, chlorinated rubber, fluorocarbon, acrylate, nitrocellulose, cellulose butyrate coatings; porcelain enamel and precoated metals, are shown.

The Study of the Passivity of Metals in Inhibitor Solutions Using Radioactive Tracers. Part I. The Action of Neutral Chromates on Iron and Steel. $KrCrO_4$ labelled with ^{51}Cr was used and log growth of an oxide film similar to that in air was demonstrated. An adsorbed layer containing Cr is also shown to be present. Inhibition persists as long as chromate ion is sufficient to sustain growth of the oxide film.

Roto-Finish and Corrosion Control. D. J. Fishlock. (*Corros. Techn.*, 1958, **5**, Sept., 287–290). A description is given of the Atram range of phosphating solutions and of various metal-cleaning techniques introduced from Continental practice by Roto-Finish Ltd.

Volvo Mechanises Its Phosphating Line. (*Steel*, 1958, **143**, Aug. 25, 100–101). The phosphating machine at a Swedish car-body plant is described. Bodies are conveyed through the plant at the rate of 33 per h. Operations are: degreasing, water rinsing, phosphating, water rinsing, dilute chromate rinsing, drying, and application of air-drying synthetic primer.—D. L. C. F.

Sacrificial Metals as a Base for Organic Coatings. K. Tator. (*Ind. Eng. Chem.*, 1958, **50**, Aug., 77A–78A). Galvanized or Al-coated steel protected with vinyl or other synthetic resin coatings show good resistance to weathering.

Plastic Linings and Coatings for Steel Water Pipe. G. E. Burnett and C. E. Selander. (*J. Amer. Water Works Assoc.*, 1958, **50**, Aug., 1065–1075). Types of plastics are briefly described. (32 refs.).

Protecting Steelwork with Epoxide Enamel. G. A. Dell. (*Corros. Techn.*, 1958, **5**, Sept., 291–292). A description of the properties of epoxy resins cured with a polyamide-type resin, with the characteristics of the resultant enamel. Particular reference is made to one proprietary epoxide enamel.

Sheathing of Electrical Conductors, Cables, Iron and Steel Wire with Plastics. U. Finck. (*Wire*, 1958, June, 36–38). The principles of operation of screw extrusion presses for the sheathing of wire and stranded products are explained and typical machines are described.

PROPERTIES AND TESTS

New Testing Method for Supplies of Steel for Gun Barrels. E. Almc. (*Tekn. Ukeblad*, 1958, **105**, Nov. 20, 1024–1025). The NATO agreement for standardization of testing methods is compared with the 1937 Norwegian specification for testing gun barrel steel.

Contribution to the Study of the Damping and Ageing of Iron–Carbon–Phosphorus Alloys. W. Dickenscheid and H. J. Seemann. (*Rev. Mét.*, 1958, **55**, Sept., 872–875). An investigation of the influence of C and P contents on the damping characteristics of iron is described, for C contents < 0·02 and P contents between 0·015 and 0·17%.

Contribution to the Study of the Improvement in the Ductility of Low-Alloy Steels. G. Delbart and F. Maratray. (*Rev. Mét.*, 1958, **55**, Sept., 876–888). An examination of two casts containing 0·8–1·0% each of Cr, Mn and Si, both low in carbon, but containing 0·016 and 0·006%S respectively, showed that the ductility of the latter cast was superior. Additions of 0·20%Mo to each cast reduced the difference in ductility without removing it entirely.

Boron-treated Cast Steels. W. J. Jackson. (*JISI*, 1959, **192**, Aug., 350-361). [This issue].

The Process of Relaxation at a High Temperature. C. Boulanger. (*Rev. Mét.*, 1958, **55**, Sept., 889–898). A description is given of tests, using a specially-designed hysteresis apparatus, in which the modulus of elasticity and internal friction of various metals and alloys are investigated as a function of temperature.

The Flow of Impurities to an Edge Dislocation. R. Bullough and R. C. Newman. (*Proc. Roy. Soc.*, 1959, **249**, Jan. 13, 427–440). A calculation of the number arriving as a function of time is presented. Experiments on the strain-ageing of steels were in excellent agreement.

High-Temperature Cryometry. G. Petit. (**Bull. Soc. Chimique de France*, 1956, 230–243). The theory of the extension of the ordinary solution and dilution laws to molten systems is given, and experiments on various systems resembling slags and the apparatus used are described. The solution of Al_2O_3 in cryolite is worked out in view of its electrochemical importance.

Activity Coefficients and Chemical Potentials of Solute Elements in Molten Iron. E. Horikawa. (*Tetsu to Hagane*, 1958, **44**, May, 533–541). The lattice model of liquids is applied to molten iron containing foreign elements. Theoretical expressions for chemical potential and activity coeff. are derived, using a method analogous to that of Bragg and Williams. These apply to the cases of three interstitial elements, three substitutional elements, and two elements of each type. (12 references).—K. E. J.

The Solubility of Carbon in Alpha-Iron. E. F. Petrova, M. J. Lapshina and L. A. Schvartsman. (*Doklady, A.N.*, 1958, **121**, (6), 1021–1024). The method and apparatus for the determination of the quantities of carbon in α-iron are described. It is based on the measurement of partial pressures of CO and CO_2 in the gases circulating over the sample of iron and calculating the equilibrium constant. The results of the determination of the solubility of carbon in α-Fe below and above the eutectoid temperature are given and discussed. The data obtained can also be used for the calculation of the amount of carbon in α-Fe in equilibrium with austenite and for the exact determination of the boundary law between austenite and ferrite.—S. I. T.

The Role of Iron as an Acceptor in an Iron–Carbon Alloy. I. N. Frantsevich, D. F. Kalinovich, I. I. Kovenskii and V. V. Pen'kovskii. (*Doklady A.N.*, 1958, **121**, (2), 277–279). The study of strength of bonds as conditioned by a donor-acceptor factor helps to elucidate the heat-resisting properties of Fe–C alloys. The present experiments consisted of annealing carburized iron wires covered in their central part with ^{59}Fe, at 900°–1100° C during periods of 12–40 h. It was found that the carbon atoms become ionized and migrate to the cathode, and the electrons thus liberated fill the vacant places in the d-level of iron atoms, which move towards the anode. The displacement of the active zone, the velocity of the movement of the zone and other numerical data, are given in tabular form.—S. I. T.

Solubility of Carbon in Ferrosilicomanganese. P. V. Geld and M. S. Petrushevskii. (*Doklady A.N.*, 1958, **120**, (1), 144–147). The solubility of carbon in Fe–Mn–Si alloys containing up to 50% of Si at 1460° C in the atmosphere of argon was determined, and a set of empirical equations was calculated and various diagrams constructed.—S. I. T.

Activity of Carbon in Liquid Iron Alloys. J. Chipman and T. Fuwa. (*Tetsu to Hagane*, 1958, **44**, Mar., 215).—K. E. J.

Effect of Silicon on the Solubility of Oxygen in Iron and Chromium Melts. V. V. Averin and A. M. Samarin. (*Doklady A.N.*, 1958, **120**, (6), 1253–1254). The influence of up to 1·5%Si on the solubility of O_2 in liquid melts of Fe and Cr at 1600° C was studied. It is concluded that the addition of Si lowers the solubility of O_2 while the addition of Ni does not affect it.—S. I. T.

Mn-S Equilibrium in C-Saturated Molten Iron. IV. On the Desulphurization Product. M. Inoue. (*Tetsu to Hagane*, 1958, **44**, Mar., 219–221). The apparatus includes a graphite crucible and furnace. Relationships are shown between the Mn contents in a slag containing Fe, Mn and S and in the metal, the Mn and S in the metal, and the MnS content in the slag and Mn content in the metal. K. E. J.

Equilibrium Constant of Reaction. $Al_2O_3(s) = 2Al + 3O$. T. Araki. (*Tetsu to Hagane*, 1958, **44**, Mar., 217–219).—K. E. J.

Temperature Control and Strain Measurement in Tensile Tests at High Rates of Heating and Loading. B. F. Billing and M. S. Binning. (*TIDU* 6014, List 971, pp. 16). Apparatus is described for both manual and automatic control of the temperature of a specimen heated to 1200° in less than 30 sec by heavy a.c. and held for up to 30 min. An extensometer for use up to 1100° is described as well as a method of measurement and recording of strain enabling a 0·2% proof stress to be reached in about 10 sec.

Comments on the Results of Tensile Tests and Fatigue Tests on Butt-Welded Joints in Alloy Steel Tubing. M. S. Fisher. (*TIDU* 5715, List 970, pp. 13). Welds in drawn and blued tubing had fatigue limit values between 11·8 and 9·0 tons/in² irrespective of composition and tensile strength of materials or joint. Hydrogen-arc welds were slightly stronger than gas welds, but the difference may not be significant. Metallic arc welds are stronger in tension, but much weaker in fatigue. Flash welds gave the highest and most consistent results. Simple heat treatments gave no improvement. Gas-welded repairs of broken or faulty joints were satisfactory in tension, but tended to be low in fatigue strength.

Study on Stress-Rupture Characteristics of Tubular Specimens. H. Maruoka. (*Tetsu to Hagane*, 1958, **44**, Mar., 424–426). The test machine is described, and results are given for tubes in 18 Ni–10Cr–Nb and 18Ni–10Cr–Ti steel.—K. E. J.

Material Rating Based upon True Stress/Strain Properties. J. Marin and M. G. Sharma. (*Weld. J.*, 1958, **37**, Aug., 375s–378s). The authors present a study of material ratings based upon the nominal and true mechanical properties in tension. The results show that these two ratings are not the same.—T. G.

Determination of the Transition Temperature and the Notch Brittleness of Steels by Static Tensile Tests. C. Pfeiffer. (*Jernkont. Ann.*, 1958, **142**, (11), 720–730). A test in which the test-piece should have a width of at least 1½ times the plate thickness, but otherwise is variable is described which is better related to practical conditions than a test-bar of fixed size. A cold-pressed notch of small bottom radius is used and in some cases the bar was afterwards tempered. Tests on non-ageing ship plate showed that the pressing was enough to start brittle fracture. Two methods of finding the temperature at which brittleness appears are given, one being measurement of lateral contraction.

Studies on the Bending Test of Forged Steels. III. Influence of Non-Metallic Inclusions on Bendability. IV. Influence of Micro-Structure on the Bendability. V. Influence of Hydrogen on Bendability. VI. Distribution of Plastic Strain on the Bent Surface. S. Onodera. (*Tetsu to Hagane*, 1958, **44**, Mar., 397–400; 400–401; 401–403; 403–404). III. Correlations are given between bend test results and distribution ratio and thickness of large inclusions, and also mean cleanliness tested by various Gakushin methods. IV. The relationship between bend test results and pearlitic grain size is shown. For good, indifferent and poor results in the bend test, the Widmanstätten structure and tensile properties are shown. V. Properties of various specimens from a 0·4%C steel forging are compared with hydrogen segregation. VI. Results are given for plain C and Ni–Cr–Mo steels.—K. E. J.

Studies on Shock Resistance of Steel. II. Some Experiments on W–Co, Cr–Ni–Mo and Si–Mn–Mo Tool Steels. M. Tanaka. (*Nippon Kinzoku*, 1956, **20**, Nov., 626–630). Some results of static and dynamic bending tests and of deformation tests are given for specific commercial steels, and these are compared for shock-resistance and allied properties.—K. E. J.

The Influence of Heat-Treatment Schedules on the Anisotropy of the Properties of Tool Steel Sheet. E. I. Astrov. (*Metall. Obra. Met.*, 1958, (9), 33–38). Steels U8GA and 85KhF were used in the form of cold-rolled sheet and strip and bend-tests were carried out as well as tensile and impact tests in the direction of rolling and at 45° and 90° to this. Magnetic anisotropy was also measured. Above the critical normalization temperature the anisotropy due to fibre structure fell to one-third and crystallographic anisotropy to about one-half. After hardening in oil and tempering and after stepwise and isothermal hardening the anisotropy was that of ordinary annealed or normalized steel. Inclusions increased the amount of anisotropy.

On the Plastic Bending Test. I. K. Sato. (*Tetsu to Hagane*, 1958, **44**, Mar., 405–407).

Study on the Hot-Ductility of Steels. I. Study on the Hot Torsion Test. T. Morishima. (*Tetsu to Hagane*, 1958, **44**, May, 552–559). The test apparatus is described and shown to have fairly good reproducibility and accuracy. It is more useful than high-temperature tensile or impact tests. Some results are given for carbon and alloy steels, and for the effect of MnS on hot ductility. (12 references).

Determination of the Relationship between the Shear Moduli of Pure Iron and Ferrous Oxide at 20° C. M. Cagnet, G. Collette and J. Moreau. (*Compt. Rend.*, 1958, **247**, Dec. 1, 2002–2003). Internal friction measurements in torsion by the compound pendulum method were made on iron oxidized symmetrically at 880° C. There was some doubt about the thickness of the oxide layers but the shear modulus of the oxide was just about half that of the iron.

A Study of Impact Fatigue Strength. N. N. Davidenkov and E. I. Belyareva. (*Metall. Obra. Met.*, 1958, (9), 12–15). The effect of specimen size was investigated on 45Kh and 30KhGSA steels after tempering at 200° or at 600° C. Values of the fatigue limit corres-

ponding to 100 000 cycles were calculated. The size effect above does not account for differences between static and dynamic fatigue tests. The concept of a "dynamic effect" relating the static value to rate of load application is developed and this changes 17% in passing from high to low temperature tempering.

The Influence of Ferrite Banding on the Impact Properties of Mild Steel. W. S. Owen, M. Cohen, and B. L. Averbach. (*Weld J.*, 1958, **37**, Aug., 368s–374s). The Charpy properties of a severely banded ferrite–pearlite structure were compared with those of a random ferrite–pearlite distribution. Both types of specimens were prepared from the same steel by suitable homogenizing treatments. The results show, *inter alia*, that the brittle fracture properties are not affected by the extent of ferrite-pearlite lamination nor do they show any anisotropy even in severely banded plates.—T. G.

Effect of Sensitizing Treatment on the Low-Temperature Impact Value of Austenitic Stainless Steels. Y. Fukase. (*Tetsu to Hagane*, 1958, **44**, Mar., 344–346).—K. E. J.

Determination of Notch-Toughness by Means of a Small Specimen. V. Komárek. (*Acta Technica*, 1958, **3**, (1), 26–57). A new form of small specimen for notched bar impact testing is described, and the results are assessed.

Oxygen and the Strain Ageing of High-Purity Iron. W. Janiche and H. Beck. (*Arch. Eisenh.*, 1958, **29**, Oct., 643–651). Experiments showed that O_2 influences the hardness of high-purity iron in a similar way to C and N. The temperature of the maximum hardness value with pure iron melts containing O_2 was substantially higher than in alloys containing only C and N.

I.I.W. Commission XIII. Fatigue Tests. Monographs on Fatigue Failures. Report on Fatigue Failures in Fatigue Machines. W. H. Munse. (*Soud. Techn. Connexes*, 1958, **12**, March-April, 148–151). The author discusses fatigue failures in fatigue testing machines. Failures occurred after about 500 000 000 cycles on flexural parts, starting in fillet welds. Residual stresses in the weld may be responsible and small cracks in the weld bead may serve to initiate failure.—R. P.

The Fatigue Strength of Weld Metal. O. Eiro. (*Acta Polytech. Scand.* 1958, *Mechanical Engineering Series No.* 3, pp. 26). The fatigue strength of metal from various types of electrode, and from the weld metal from manual and automatic welds was measured, and the results discussed.

A Study of the Fatigue Resistance of Butt Joints Welded Automatically by the 'Fusarc' and 'Unionmelt' Processes. Z. Garcia Martin and F. Cacho Falco. (*Inst. Hierro Acero*, 1958, **11**, July-Sept., 185–201). [In Spanish]. The mechanical properties, quality, structure, contraction, distortion and residual stresses of these welded joints have been studied and compared with those made by hand. Plates, 12·5 mm thick, of a 0·27% carbon steel were used in the tests. Both automatic processes gave the same fatigue results (pulsating tension) though more scattered values were obtained with Fusarc welds. Both gave better results than the manual welds did, this is thought to be due to differences in the materials rather than the welding technique.—P. S.

Influence of Dimensions on the Fatigue of a Notched Specimen. P. Kuhn. (*Rev. Mét.*, 1958, **55**, Sept., 860–863; discussion 863–864). Using Heuber's formula (*Kerbspannungslehre*, Springer, 1937) a curve is derived for calculating the fatigue limit of a notched specimen, given the materials constant and the fatigue limit of the unnotched specimen.

The Effect of Internal Friction on the Fatigue Failure of Cyclically Deformable Machine Parts. Y. S. Postrigach and M. I. Chaevskii. (*Doklady A.N.*, 1958, **121**, (2), 268–270). This is a study of the influence of temperature, developed through torsional deformation of steel parts, on fatigue failure. Sets of equations and curves are presented. It is concluded that at stresses above the limits of fatigue, the effect of internal friction is markedly on the increase and can lead to thermal stresses of values exceeding those calculated by the authors. Parts subjected to high cyclic loads should be evaluated taking into consideration the absorbing properties (i.e. internal friction) of the material.—S. I. T.

Influence of the Lamellar Pearlite Structure on the Transition Temperature and Fatigue Strength of Carbon Steels. S. Tsujimoto. (*Tetsu to Hagane*, 1958, **44**, Mar., 394–395).

Stress-Relaxation Tests on Welded Specimens of Consett "Low-Tem" Steel. A. J. Fenner and A. Cowan. (*U.K. Atomic Energy Authority, Industrial Group, Tech. Note, No.* 113, 1958, pp. 11). The tests were made with special reference to internal stresses in the pressure shell of a thermal reactor. Appreciable relaxation was observed at all temperatures in the range 350–400°, but the rate was fairly low in weld metal. At the normal temperature for stress relief a relatively short time would be necessary to reduce stresses to a low value.

Low Ductility in Cast Carbon Steel. T. A. Cosh and W. J. Jackson. (*Metallurgia*, 1958, **58**, Oct., 190–192). Investigations by B.S.C.R.A. showed that failure to attain high ductility in plain carbon steel castings was, in the cases examined, attributable to inclusions. The nature and cause of these inclusions are discussed, and methods of minimizing their formation outlined.

Die Tests Elongation. (*Steel*, 1958, **143**, Sept. 8, 101). A testing device is described which consists of a set of V-shaped dies. These produce progressively greater elongations along a specimen and the point of fracture indicates forming limits.—A. G.

Large Range Pneumatic Extensometer. N. L. Svensson. (*Chartered Mech. Eng.*, 1958, **5**, Nov. 420–422). An extensometer for measurement of strains of elastic order after pre-strains of up to 20% with a 1 in. specimen, requiring measurement of movements of $< 10^{-3}$ in. over a range of $\pm$ 0·2 in. is described using a pneumatic gauging unit.

Flexural Behaviour of Mild Steel Beams. K. Farnell. (*Engineering*, 1958, **186**, Nov. 21, 674). Increases in load capacity produced by initial tensile overstrain are considered.

Investigation of Steel for Flat Wood-Working Blades. A. G. Ivanov, E. V. Kruchinina and V. S. Rybalko. (*Stal' Supplt. Metallogr. and Heat Treatmt.*, 1958, 256–274). The investigations determined the critical transformation intervals, the mechanical properties as given by bending torsion tests, and the cutting qualities of the finished tools, which characterize the numerical value of the wear of the edges of the blades, the variations in the micro-geometry of the blade, the purity of the surface obtainable and also the expenditure of power in wood-working.—L. H.

Notched-Bar Tests on Steel. P. Kövesi. (*Iron Coal Trades Rev.*, 1958, **177**, Oct. 10, 858). A letter, reproducing a diagram from the Stahleisen-Kolender, 1958.

Prediction of Transition Temperature in a Notched-Bar Impact Test. J. A. Hendrickson, D. S. Wood and D. S. Clark. (*A.S.M. preprint*, 96, 1958, pp, 13). Previous ideas of the mechanism of initiation of brittle fracture in mild steel have been applied. The predicted transition temperature agrees well with the experimental values.—E. E. W.

Effect of Thermal and Mechanical Treatments at Forging on the Impact Strength of Structural and Alloy Steels. E. N. Sokolkov, L. V. Smirnov and S. N. Petrova. (*Fiz. Met.*, 1958, **6**, (2), 276–280). It has been experimentally confirmed that the heat- and mechanical treatments during forging and rolling weaken the tendency of such steels to reverse temper brittleness. This is due to the localization of the deformation along the grain boundaries of the original austenite grains, owing to the distortion of the crystal lattice in the intermediate zones. This effect can also be observed in other treatments, for example, pressing or stamping, because the recrystallization of the cold-hardened austenite is then checked.—L. H.

The Effect of Strain and Temper Embrittlement on the Impact Transition Temperature of an AISI E-6150 Steel. E. F. Steeb and P. C. Rosenthal. (*Trans. Met. A.I.M.E.*, 1958, **212**, Oct., 706–707). It was found that this steel was rather highly susceptible to temper embrittlement and the treatment described caused a rise in transition temperature of 165° F at 0% strain. Post-straining at 900° F produced an appreciable increase in transition temperature. A full record of experimental results is given.

Effect of Heat Treating on Notch Sensitivity of Chrome–Molybdenum–Vanadium Steel. Z. N. Petropavlovskaya. (*Metallov. Obra. Met.*, 1958, (7), 54–58). Notch sensitivity can be lowered and embrittlement retarded by double normalizing (1040°/960° C) and tempering at 680°.

Determination of Impact Hardness and Characteristics of Steel. B. A. Vandyshev and F. S. Savitskii. (*Fiz. Met.*, 1958, **6**, (3), 534–539). The test material consisted of 6 types of steel in round bars containing 0·17%; 0·22%; 0·47%; 0·63%; 1·06% and 1·16%C. Findings show that the dynamical coefficient depends on the carbon content and the greatest source of error in determining static hardness by impact instruments is precisely this dependence on the carbon content of steel. The hardness value will characterize the resistance to plastic deformation of a steel undistorted by recrystallization. This is important to know for the hot-working of metals.—L. H.

The Effect of Neutron Irradiation on the Impact Properties of En2 Mild Steel. I. L. Mogford, A. T. Churchman and D. Hull. (*AERE M/R* 2485, *U.K. Atomic Energy Authority, Research Group Report*, 1958, pp. 9). After a dose of $9{\cdot}3 \times 10^{18}$ neutrons/cm² the increase in transition temperature was found to be nearly proportional to (dose)$^{\frac{1}{2}}$. At the maximum irradiation level the risy was 75° C. There is a reduction in the energy absorbed in ductile fracture after irradiation of about 50%.

Behaviour of Certain Alloys Subjected to Dynamic Loading. R. G. Crum and F. T. Mavis. (*A.S.T.M. Bull.*, 1958, July, 88–91). An impact apparatus for $\frac{1}{4}$ in. diameter specimens is described with cathode ray recording and data on steels and other metals are shown with comparisons with static loading.

Explosive Press for Use in Impulsive Loading Studies. E. W. LaRocca and J. Pearson. (*Rev. Sci. Inst.*, 1958, **29**, Oct., 848–851). Single and double-acting presses producing pressures estimated at several million psi are described.

Metallographic Studies of Metals after Explosive Shock. C. S. Smith. (*Trans. Met. A.I.M.E.*, 1958, **212**, Oct., 574–589). Fe is hardened by a shock of ~ 400 kbar more than by cold-rolling to 95% reduction, although the dimensions of the sample are virtually unchanged. Neumann bands are developed on compression. Above 130 k bars, transition to a different phase occurs, which after return to atmospheric pressure leaves a complex microstructure. It is shown how the structure can be used to measure shock velocities.

Application of the Griffith-Irwin Theory of Crack Propagation to the Bursting Behaviour of Disks, Including Analytical and Experimental Studies. D. H. Winne and B. M. Wundt. (*Trans. A.S.M.E.*, 1958, **80**, Nov., 1643–1658; *A.S.M.E.*, 1957, 57-A-249, pp. 23). Tests on large bored notched discs removed from long rotor forgings are described. Average tangential bursting stress was as low as 26% of yield strength for brittle materials and as high as 92% for those approaching ductile behaviour. Fracture toughness is calculated from the theory and agrees with slow notched-bend tests and appears to be independent of size of specimen.

Influence of Texture Deformation on the Mechanical Properties of Metals. M. L. Bernshtein and M. L. Grinberg. (*Stal' Supplmt. Metallogr. and Heat Treatmnt.*, 1958, 65–78). Tests are described to investigate the dependence of mechanical properties of cold-deformed Armco iron on the type of structure present during different stress

conditions, the findings show that a polycrystalline texture can result in an increase of the mechanical properties because of the anisotropic orientation and distribution of the cold-hardened zones.—L. H.

Mechanical Properties of Deformed Metastable Austenitic Ultra High Strength Steel. D. J. Schmatz and V. F. Zackay. (*A.S.M. preprint* 110, 1958, pp. 13). A new hardening technique for steels containing 0·28–0·57%C has been tested which consists in plastically deforming at a sub-critical temperature. Several suggestions are advanced to account for the mechanisms responsible for the improvements in mechanical properties.—E. E. W.

Fatigue of Metals Caused by a Relatively Few Cycles of High Load or Strain Amplitude. P. P. Benham. (*Met. Rev.*, 1958, **3**, (11), 203–234). A review of studies on the causes of failure of metals under relatively few repeated loads (up to $\sim 10^5$ cycles). Changes in properties, energy change and change in ductility during cycling, effect of prior cycling on the stress-strain curve to fracture, and various explanations are referred to. A summary of data is added. (51 refs.).

An Introduction to Fatigue in Marine Engineering. B. K. Batten. (*Trans. Inst. Marine Eng.*, 1958, **60**, Nov., 331–355). Methods of crack detection are considered and the question of stress concentration as affected by design is reviewed in detail. In an appendix types of crack are shown and interpretation of fracture pattern is dealt with. Fretting corrosion is included and numerous failures in service are discussed. An attempt to summarize design factors is made.

Extra Fatigue Strength is Now Available. W. S. Hyler. (*S.A.E.J.*, 1958, **66**, Oct., 40–41). Machining and mechanical polishing can increase fatigue strength considerably. Comparisons with electropolished specimens are reported and stress and depth of hardening are estimated.

Fatigue Testing of Vehicle Components. T. C. F. Stott. (*Inst. Mech. Eng., Automobile Div. preprint*, 1958, pp. 10).

Thermal Stresses in Design. Part 3: Basic Concepts of Fatigue in Ductile Materials. S. S. Manson. (*Mach. Design*, 1958, **30**, Aug. 7, 100–107). Stress-strain mechanisms and failure by thermal cycling are discussed.

Thermal Stresses in Design. 5. Interpretation of Fatigue Data for Ductile Materials. S. S. Manson. (*Mach. Design*, 1958, **30**, Sept. 4, 126–133). Effects of thermal stress cycling are discussed, with pre-strain effects, comparisons with mechanical fatigue and applications to design.

Electro-Magnetic Vibrators: Use in Creep and Fatigue Tests. (*Iron Steel*, 1958, **31**, Oct., 504–506). The specially-designed test equipment used by B.I.S.R.A. and the Clarendon Laboratory to study fatigue at low temperatures and the effect of combined creep and fatigue are described, and results obtained are discussed.

The Influence of Shot-peening on the Fatigue Strength of Metals (1st Report). S. Takeuchi and T. Honma. (*Nippon Kinzoku*, 1958, **22**, Jan., 14–18). In preliminary rotary bend tests the endurance limit of pure iron, Si–Mn steel, 17SAl alloy, and 40/60 brass was increased by shot peening several times compared with that of polished specimens; the fatigue limit was increased by 3–13% as the result of shot-peening and reduced by annealing after shot-peening. (**2nd Report**); **On the Effect for Decarburized Steels,** (18–22). Decarburization of Si–Mn steel caused a 50% drop of the fatigue limit, compared with polished steel. Shot peening had the opposite effect. (**3rd Report**). **On the Effect for Notched Steels.** (Feb., 55–59). In investigations of Si–Mn spring steels, heat treated to $140 \sim 150$ kg/mm² tensile strength, it was found that shot peening was more effective on notched than on polished specimens, even when the shot could not reach the root of the notch. (**4th Report**) **On Its Effect on Corrosion Fatigue.** (Feb., 59–63). The effect of shot-peening on the fatigue strength of Si–Mn steels was investigated in the corrosive medium of distilled water and of diluted sulphuric acid. Though attenuated, the beneficial effect of surface stressing was still manifest.

The Problem of Fatigue Relative to the Optimum Useful Characteristics of Steels. M. M. Alvaro Periane. (*Tecn. Met.*, 1958, **14**, Mar.-Apr., 39–47). [In Spanish]. The author suggests that it is possible to use the normal Spanish steels at higher rupture strengths in spite of the appreciable loss in ductility. The importance of fatigue behaviour is discussed and results of tests made to examine the possible relationships between the static and fatigue characteristics are given. These results show that the fatigue limit, even in specimens with theoretical stress-concentration factors of 1·8, has increased with the rupture strength in spite of the loss in toughness. Low impact figures (5 kg/cm²) do not form an obstacle to the use of the two steels examined. (F.123 and F.127 Ni–Cr and Ni–Cr–Mo high strength Spanish aviation steels).—P. S.

A Compilation and Interpretation of Cyclic Strain Fatigue Tests on Metals. J. F. Tavernelli and L. F. Coffin, jun. (*A.S.M. preprint*, 112, 1958, pp. 13). From consideration of accumulated data there is proof of the validity that, for all metals tested, the plot of the log of the plastic strain range against the log of the number of cycles to failure is a straight line of slope equal to $-\frac{1}{2}$. It is also reasonable to plot this type of fatigue curve from the fracture ductility of a metal or alloy in a simple tensile test.—E. E. W.

Fretting-Fatigue Failure of Titanium–Steel and Steel–Steel Joints. W. L. Starkey, S. M. Marco and J. A. Collins. (*Prod. Eng.*, 1958, **29**, Sept., F15–F17; from an A.S.M.E. paper). Results of tests to evaluate the fatigue damage caused by fretting Ti angainst steel indicate that systems of stress concentration and other effects caused by pit digging, asperity contacts and cyclic nominal stress field in the part, combine to initiate and propagate cracks. Cold rolling, shot peening and other measures can produce an optimum residual-stress field which mitigates stress propagation.

A Mechanism for Nonpropagating Fatigue Cracks. L. F. Coffin jun. (*A.S.T.M. preprint*, 1958, (72), pp. 6). Fatigue cracks at the base of sharp notches are non-propagating. A mechanism is suggested by which the stress and strain at the base of the notch are fully reversed in cyclic load, but that full reversal does not occur for a crack because of closure. The effect of superimposed mean stress is correctly predicted.

Inclusions: Nuclei for Fatigue Cracks? (*Metal Progress*, 1958, **74**, Aug., 120–122). An account is given of an A.S.T.M. meeting at which the effect of inclusions on fatigue life of steel was discussed. The medium and long-life fatigue strength of certain steels decreases substantially for each 0·001 in. increase in the size of the inclusions, particularly at the high hardness levels; the effect is small with hardness levels below 140 000 p.s.i.; the effect of inclusions below 0·00025 in. size may be insignificant. One steel made in a series free from silicate inclusions gave the highest fatigue strength ever reached by any metal. Torsional fatigue life of cold drawn music wire is decreased by the presence of inclusions.—D. L. C. F.

Tests Demand Constant Temperatures. (*Steel*, 1958, **143**, Sept. 1, 77). Creep and stress-rupture tests are described in which thermocouples are so located as to secure temperature control.

A Simple Creep Strain Recorder for Use with Wire Specimens. J. Congleton and R. N. Parkins. (*J. Sci. Instr.*, 1958, **35**, Oct., 379–380).

How to Simplify Tests for Primary Creep Data. B. B. Muvdi and C. J. Giemza. (*S.A.E.J.*, 1958, **66**, Aug., 46–48). A graphical method of obtaining creep curves from two tests at any one temperature and the construction of master diagrams for estimating creep between test temperatures is given. 17–7 PH stainless steel is included.

Investigating the Influence of Preliminary Cold Deformation on Creep. G. I. Nosova and V. M. Rosenberg. (*Fiz. Met.*, 1958, **6**, (2), 321–325). The alloys Fe–Ni–Cr–Co with varying Co content were subjected to bending tests at various temperatures. Results obtained are considered only qualitatively and not as absolute values. At temperatures below 600° the main factor of deformation is the structure. At higher temperatures when the structure which corresponds to a high resistance can no longer persist, the main factor becomes the chemical composition. The rate of creep for specimens hardened by cold deformation, but not previously stabilized increases rapidly at all recrystallization temperatures under the influence of temperature and stress, and the softening of specimens which were previously deformed proceeds to a larger extent than it does under the influence of temperature alone.—L. H.

The Application of Time Temperature Parameters to Accelerated Creep-Rupture Testing. S. S. Manson, G. Succop and W. F. Brown, jun. (*A.S.M. preprint*, 106, 1958, pp. 15). A procedure is described in which parameter constants and the master curves were determined over a wide stress range. The results are compared with isothermal data established by conventional procedures and show good agreement.—E. E. W.

The Phenomenon of "Negative Creep" in Alloys. R. W. Fountain and M. Korchynsky. (*A.S.M. preprint* 91, 1958, pp. 13). It is shown by the use of some Fe–W and Ni–Mo alloys that the amount of contraction can be calculated from the known phase relationships and crystallographic data. Alloys showing strong tendencies to negative creep may present difficulties during annealing treatments after prior cold work because of additional build-up of internal stresses, which could cause cracking.—E. E. W.

Generalized Creep Criteria Diagram Based on New Relationships Among Stress, Creep Rate and Service Life. I. A. Oding and V. S. Ivanova. (*Issledovaniya Po Zharoprochnym Splavam*, 1956, **1**, 52–59). The most accurate values for rupture strength are obtained by extrapolation of short-period test results on semi-log paper, using an equation given. The ductility reserve is also important and the creep diagram includes this.

Prediction of Creep in Bending from Tension and Compression Creep Data when Creep Coefficients are Unequal. W. N. Findley, J. J. Poczatek and P. N. Mathur. (*Trans. A.S.M.E.*, 1958, **80**, Aug., 1294–1298). Tests on non-metallic materials gave general support to the theory, but are insufficient for an accurate check.

Estimation of Ductility (Creep at Rupture) of Creep-Resisting [Steels and] Alloys at High Temperature. A. V. Stanyukovich. (*Zavods. Lab.*, 1957, **23**, (4), 476–484). Creep at rupture is defined as total creep at rupture less initial elastic and plastic deformation. An empirical "mean creep rate" is also defined. Values are found for various steels and the stages of creep are discussed and formulated. The embrittlement range is reviewed and preliminary attempts to establish relationships between the various factors are made.

Some Creep and Tensile Properties of a Corrosion Resisting Steel to British Standards Specification S.80, at 22° C, 250° C, 400° C and 550° C. R. A. Whittaker. (*Royal Aircraft Establishment Tech. Note Met.* 280, 1957, Dec., pp. 17). Constant load creep and short term tensile tests are reported. At the two lower temperatures, the curves are not of normal type; a rapid increase in strain is followed by a small creep rate. The effect of 0·1% prestrain at room temperature before testing at 250° C is to reduce the amount of strain on loading by $\sim 60\%$. Engineering data are derived from the creep curves at 400° and 500° C.

Solve Lamellar Phase Problem in A-286. K. Metcalfe. (*Iron Age*, 1958, **182**, July 3, 72–74). The answers to the problem of stress-rupture failure of an austenitic stainless steel used for high temperature bolts have been deduced. Trouble arises from the formation of a lamellar precipitate that is similar in appearance to the pearlite found in carbon steel. The trouble can be avoided: (*a*) in parts that are not processed by cold working, by keeping the boron content above 0·001%; (*b*) where severe cold-working between solution-

treating and ageing is necessary, by re-solution treating at 2000° F.—D. L. C. F.

A Note on the Comparative Buckling Properties of Mild-Steel and Aluminium-Alloy Plating. G. R. Smith. (*Shipbuilder*, 1958, **65**, Sept., 533–537). The work of Montgomerie and of Muckle is used to set up corresponding figures for scantlings with the same factor of safety against buckling.

Dispersion in Jominy Testing. (*Centre Doc. Sidér. Circ.*, 1958, **15**, (5), 1079–1083). The proceedings of a conference on Jominy Quench Testing are reported. The reproducibility of the test, the influence of oil and water quenching and the calculation of Jominy curves are among the subjects discussed.—B. G. B.

On the Abnormal Shapes Appearing in Jominy Curves of Low Alloy Steels. T. Nemoto. (*Nippon Kinzoku*, 1956, **20**, May, 243–247). Quenching diagrams were determined for steels containing 0·5%Mo, 1·0%Cr and 0·5%Mo, 1·0%Cr and 1·0%Mo, and 2·0%Cr and 1·0%Mo, respectively. Hardenability increases as Cr and Mo increase. The transformation rate in the Ar′ "nose" is larger than that in Ar_1 "nose". The amount of retained austenite is a maximum in the range of cooling rate where abnormal shapes appear in the Jominy hardenability curves.—K. E. J.

Multiplying Factors for the Calculation of Hardenability of Hypereutectoid Steels Hardened from 1700° F C. F. Jatczak and D. J. Girardi. (*A.S.M. preprint* 99, 1958, pp. 15). The hardenability effects of 1% carbon steels containing up to about 1·5%Mn, Si, Cr, and Ni and 1·0%Mo, singly or in combination have been determined.—E. E. W.

The Secondary Hardening of Structural Steels. L. I. Kogan and R. I. Entin. (*Fiz. Met.*, 1957, **5**, (2), 349–354). A relationship was established between the carbon content of residual austenite and the intensity of the effect of secondary hardening.

Effect of Lubricant on the Penetration of a Diamond-Pointed Indenter into Ductile Metals. V. D. Kuznetsov and A. I. Loskutov. (*Fiz. Met. Metallov.*, 1956, **2**, (3), 509–513). Brinell and Rockwell hardnesses are not appreciably affected by lubricant but cone indenter penetration at high load may be. This has been confirmed and the lubricated and non-lubricated indentation curves are shown to diverge at higher loadings. The effect is ascribed purely to the lubrication effect and not to any crystal structure loosening by the surface-active material.

Recrystallization, Structure, and Hardness of Low Carbon Steels Containing up to 1% Copper. R. L. Rickett and W. C. Leslie. (*A.S.M. preprint* 108, 1958, pp. 24). In the presence of Cu recrystallization was retarded and hardness increased after annealing. Elongated grain structure in 0·4–0·9%Cu steels may be obtained by cold-reduction followed either by heating at a rate of about 200° F/h to 1300° F or by pretreatment at 950° F and annealing. Under these conditions the hardening effect may be minimized.

Dependability of Nickel Steels. (*Prod. Eng.*, 1958, **29**, Sept., D11; from *Inco Nickel Topics*). Tables showing the hardening response of alloy steels with varying times and temperatures of heating, and different prior structures. The fullest hardening response is obtained with 3½%Ni steel.

Ageing of Steel. B. R. Nijhawan. (*Tisco*, 1958, **5**, Oct., 177–182). A review. Strain-ageing is treated by dislocation theory and quench ageing is considered, and the effects of both. (22 refs.).

A Note on the Effect of Ageing on Stress Relief. M. M. Hallett. (*Brit. Foundryman*, 1958, **51**, July, 355). A number of low-alloy high-strength grey-iron castings were tested for internal stress (two months after casting) in the as-cast condition, they were then aged for a year in outside weather conditions or in an office, and tested again. The internal stress went up from the original 3 t/in² to about 4·5 t/in² (average) after ageing.—T. G.

Study of the Ageing of Mild Steel with the Electron Microscope (1st Report). On the Behaviour of Precipitates in Ageing at 70° C. Y. Tanabe and N. Takamoto. (*Nippon Kinzoku*, 1958, **22**, Jan., 36–40). Ageing studies of 0·08%C steel comprised measurements of Rockwell Hardness and electron micrographs. The age-hardening curve exhibited two steps. The structural changes are described in detail and are tabulated in English.—J. G. W.

Nitrogen as the Alloying Element in Steel. XX. The Effect of Arsenic and Copper on Strain-Ageing in Steels. Y. Imai and T. Ishizaki. (*Nippon Kinzoku*, 1956, **20**, Dec., 673–676). Work on low-carbon steels shows that the solid solubility of N in α-iron is reduced by adding As and Cu; hence strain-ageing due to N is decreased, but that mainly due to carbon was increased by adding As and Cu.—K. E. J.

Changes in the Mechanical Properties of Steel in Storage After Thermal Treatment. K. A. Bessonov and Yu. E. Bondarev. (*Fiz. Met.*, 1958, **6**, (3), 571–573). A prolonged storage at room temperature of Type 3 steel resulted in an appreciable reduction in the uniformity of elongation, which may be explained by the cold hardening of the ferrite under the effect of stresses of the second order which occur owing to the difference in the coefficients. Linear expansion, yield point, long-term resistance and relative reduction of area increased slightly. The impact strength was reduced at all temperatures investigated and the threshold of cold shortness pushed to higher temperatures.—L. H.

Temper Embrittlement. T. B. Singh. (*Tisco*, 1958, **5**, Oct., 166–176). Detection by grain boundary etching methods is discussed and the nature of the phenomenon and its occurrence in steels of various types and the effects of alloying elements and influence on mechanical properties are reviewed.

The Nature of Brittle Failure in Steel. N. V. Tolstoguzov and A. D. Kramarov. (*Stal' Supplt. Metallogr. and Heat Treatmt*, 1958, 112–121). The investigation was carried out by etching with ethereal picric acid. This revealed the differences between the brittle and plastic condition of the steel. The initial austenite boundaries were revealed. Brittleness develops along former boundaries. In contrast the temper-brittle failure of the same steel occurs mainly along the grains. This means that temper brittleness affects the processes occurring at the boundaries of the austenite grains.—L. H.

The Ductile-Brittle Transition in the Fracture of α-Iron. I. N. J. Petch. (*Phil. Mag.*, 1958, **3**, Oct., 1089–1097). A criterion for transition from ductile to cleavage fracture is calculated for notched specimens. The temperature dependence of this transition is supposed to arise more from the Peierls-Nabarro stress required to move a free dislocation than from the temperature-dependence of the locking of a dislocation source. The temperature is a function of grain size, the friction of a free dislocation, the strength of dislocating locking, the degree of triaxiality of the applied stress and on the elastic constants. **II.** J. Heslop and N. J. Petch. (1128–1136). Agreement of the above expression is demonstrated for notched specimens with variations in grain size and with the friction of a free dislocation. A calculation is given of the effect of the dislocation locking strength on transition temperature and this is in reasonable agreement with the observed temperature for steels with locking weakened by Mn. It is suggested that strain and strain-ageing raise the transition temperature by the increased tensile stress available at fracture for crack propagation.

Ductile—Brittle Transition in Steels Irradiated with Neutrons. D. Hull and I. L. Mogford. (*Phil. Mag.*, 1958, **3**, Nov., 1213–1222). Measurements on control and irradiated specimens of En 2 steel were made at 20°, − 78° and − 196° C and the results are shown to be in agreement with Cottrell's relation for the transition. The surface relating grain size, yield stress and temperature is shown.

Low Temperature Embrittlement of Austenitic Cr–Mn–N Fe Alloys. F. W. Schaller and V. F. Zackay. (*A.S.M. preprint* 109, 1958, pp. 15). Ductility has been studied as a function of test temperature, interstitial content and strain rate. Low-temperature embrittlement was explained by using a model with strain-induced martensite as the embrittling agent. At high strain rates, localized temperature increase was suggested as the cause of an anomalous increase in reduction of area.—E. E. W.

The Brittle State in the Nickel–Chromium Alloy Steel Type F.123 I.H.A. J. Ferrar Flotats. (*Inst. Hierro Acero*, 1958, **11**, Jan.-Mar., 39–57). [In Spanish]. Many of a large series of crankshafts for aero engines made in this steel (analysis quoted: 0·28%C, 0·38%Si; 0·56Mn, 2·86%Ni, 0·70%Cr, 0·018%S, and 0·040%P) were rejected because of their failure to reach specified impact values and an investigation was commenced. A series of three-dimensional curves are given showing the relationship between time and temperature of tempering and either shock-impact values or hardness figures, for forging coefficients of one and four and for longitudinal and transverse directions. Metallographic results are also given. Some of the conclusions reached in this study are: (1) the brittleness in the crankshafts was caused by tempering below 600° C with the idea of keeping to a preferred hardness; (2) the brittle zones can be passed by higher tempering temperatures (from 550° C to 650° C impact value increases from 7·5 to 14 Kg.cm^{-2} with a hardness variation of 3 Rockwell C.); (3) tempering time has little effect on impact, compared with tempering temperature; (4) sensitization by furnace cooling from 620–550° C, range gives values of the order of 3·4 Kg.cm^{-2} in longitudinal tests; (5) brittleness could be detected metallographically by the speed of etching and the persistence of the etch after a re-polish. Cohen's reagent gave the best results, surface activation agents were of no extra help. The most distinct grain boundary etch was obtained with the 450° C temper. (28 references).—P. S.

Self-Restraint in Steel in Relation to Brittle Fracture. S. A. Main. (*Iron Steel*, 1958, **31**, Oct., 479–484, 498). The relation between self-restraint and brittle fracture in ductile metals is discussed, and deficiencies in present methods of physical testing, particularly the notched-bar impact test are explained in terms of the effects of self-restraint (15 refs.)

Effect of Isothermal Hardening on the Temper-Brittleness of 12Kh2N4A Steel. I. E. Kontorovich and B. M. Voshedchenko. (*Stal' Supplement Metallogr. and Heat Treatmt.*, 1958, 104–111). Steel with a composition 0·13%C; 0·28%Si; 0·5%Mn; 0·49%Cr; 3·67%Ni; 0·018%S and 0·015%P was tested at 150° intervals between 800 and 1250° both with isothermal cooling and with oil quenching. The hardness after tempering was taken at 100° intervals between 350 and 650°. In addition, the effect of isothermal hardening was also investigated with regard to the critical temperature of brittleness, to the austenite grain size and to the impact breaking strength.—L. H.

Micromechanism of Brittle Fracture of Steel. B. S. Kasatkin. (*Izvest. Akad. Nauk. Otdel. Tekh.*, 1958, (5), 63–68). Recent X-ray examination has shown that brittle fracture is undoubtedly associated with the prior existence of local plastic deformation.—R. S.

Effect of Phosphorus and Nickel on the Cold Shortness of Medium Carbon Steels. V. I. Svechnikov and Yu. E. Yakovchik. (*Fiz. Met.*, 1958, **6**, (3), 505–511). The test material consisted of 3 experimental melts with a content of 0·143–0·16%P and a varying Ni content from 0·0 to 2·04% and of 4 experimental melts with a varying content from 0·014 to 0·25%P. The carbon content in all cases fluctuated between 0·44 to 0·54%. Graphs are given for impact strength as a function of temperature. The findings point to the fact that the temperature at the initial stages of brittleness and at the complete transition of the steel to the brittle condition depends on the properties of the ferrite, and that plastic failure is influenced by the amount of ferrite and by the form of the cementite segregations and their distribution within the ferritic matrix.—L. H.

Effect of Heat Treatment, Straightening and Ageing on the Cold Shortness of Bessemer Rail Steel. G. I. Pogodin-Alekseev and N. S. Artemenko. (*Stal' Supplt. Metallogr. and Heat Treatmt.*, 1958, 122–131). Normalizing and improving treatments of steel reduce considerably the cold shortness of Bessemer rail steel by shifting by 80° towards lower temperatures the threshold of cold shortness. This advantage also accrues to O.H. rail steel. The straightening of double and quadruple bends causes a shift of cold shortness towards higher temperatures, on an average by 40°. The effect of ageing shows itself chiefly in a reduction of the absolute values of failure under service conditions.—L. H.

Some Data Regarding the Effect of Cold Hardening of Quenched Structures on Reverse Temper Brittleness of Steel. F. P. Stepanov. (*Fiz. Met.*, 1958, **6**, (3), 483–486). It has been shown that a cold tensile deformation to 1·5% results in an increase of impact strength and considerably reduces the tendency to temper brittleness whilst a deformation to 4% eliminates completely any tendency to temper brittleness down to − 60° C.—L. H.

The Influence of the Time of Tempering at High Heat on the Occurrence of Temper Brittleness in Steel with a Varied Original Structure. T. V. Sergievskaya. (*Stal' Supplt. Metallogr. and Heat Treatmt.*, 1958, 95–103). The original structure of 45 G2 steel contains martensite of hardness 57–60 Rc and acicular troostite of hardness 43–46 Rc. Increasing the time of tempering at high heat up to 50 h does not eliminate the differences in the properties of the tempered products. The temperature of the upper critical interval of brittleness with the structure of tempered acicular troostite is independent of the time of tempering and remains higher than for steels with a structure consisting of products resulting from martensite tempering, which products are found both in a plastic and in a brittle condition. The development of brittleness, which takes 2·3 h at 550° (brittle temperature) is reduced when increasing the time of tempering at high heat (650°) for both structures investigated.—L. H.

The Response of High-Strength Steels in the Range of 180 000 to 300 000 psi to Hydrogen Embrittlement from Cadmium Electroplating. E. P. Klier, B. B. Muvdi and G. Sachs. (*A.S.T.M. preprint*, 1958, (78), pp. 23). Seven steels were studied in sustained load tests and all were found to be embrittled in some measure. Failure is affected by the experimental conditions. In the hydrogen-bearing zone a crack is initiated and may propagate to failure through overloading. Crack development depends, in part, on composition and is minimized by reduced C or increased Si.

Nucleation of Cracks by the Intersection of Twins in α-Iron. D. Hull. (*Phil. Mag.*, 1958, **3**, Dec., 1468–1469). Studies on a 3%Si iron produced by vacuum melting, electropolished and strained are reported. At 78° K the specimens broke without measurable deformation when strained in tension along the [001] axis. It was found that {112} twins had formed. Both main and subsidiary cracks formed at the intersection of twins. Two methods of nucleation were observed and these are illustrated.

The Wear of Metals. J. F. Archard. (*Sci. Lubr.*, 1958, **10**, Nov., 16–21). Work with pin and ring machines on mild and ferritic stainless steels and non-ferrous materials and carbides is summarized, and the general theory of wear is developed and the use of tracers described.

The Coefficient of Friction of Metals. C. Rubenstein. (*Wear*, 1958, **2**, Nov., 85–96). Studies of a steel ball sliding on Indium are described and discussed.

Research into Friction. (*Engineer*, 1958, **206**, Nov. 28, 848–852). A review on the occasion of the opening of new laboratories at Chapel-en-le-Frith. Research processes are described including notes on powder metallurgy products for frictional uses.

A Crossed-Cylinders Friction Machine. J. F. Archard. (*Wear*, 1958, **2**, Aug., 21–27).

A New Method of Measuring the Temperature of Sliding Friction between Dissimilar Metals. A. S. Vagramyan. (*Vestnik Mashin.*, 1958, **38**, (7), 13–15; *Eng. Dig.*, 1958, **19**, Oct., 440).

Radioisotopes Measure Engine Wear. J. H. Deterding and J. R. B. Calow. (*Atomics Nucl. En.*, 1958, **9**, Sept., 312–315). A continuous recording system attached to a car was developed using irradiated piston rings and a scintillation counter in the oil circulation system. Wear traces and the effect of an additive are shown.

The Friction of Solids at Very High Speeds. I. Metal on Metal; II. Metal on Diamond. F. P. Bowden and E. H. Freitag. (*Proc. Roy. Soc.*, 1958, **248**, Nov. 25, 350–367).

Comparison of Wear Between One-way Friction and Reciprocating Friction (3rd Report). On the Wear of Cast Iron against Cast Iron. K. Okabayashi, S. Saito, and F. Nakamura. (*Nippon Kinzoku*, 1958, **22**, Jan., 22–25). The conclusions from this research are that wear increases in proportion to pressure, but the 'constant' proportion factor has to be adjusted to particular circumstances.—J. G. W.

Answers Found to 1000° F Wear and Galling Problems. C. H. Cannon. (*S.A.E.J.*, 1958, **66**, 86). Flame-plated Inconel X, using alumina, is briefly described.

Grey Iron Has Unique Wear-Resistance Properties. J. R. Drear. (*Foundry*, 1958, **86**, Aug., 62–63). The effects of the various phases and constituents of grey iron on its wear properties under conditions of boundary lubrication are discussed.—T. G.

Frictional Behaviour and Structure of the Surface Oxides of Steel. Y. Tamai. (*Brit. J. Appl. Phys.*, 1958, **9**, Aug., 338–339). A letter. A high-carbon steel surface oxidized by exposure under various conditions of time, temperature and humidity were examined for frictional behaviour. Small, but significant differences were observed.

An Evaluation of the Effectiveness of Penetrants. R. E. Kleint. (*Nondest. Test.*, 1958, **16**, Sept.-Oct., 421–429). Both fluorescent and dye methods were examined including proprietary preparations. The post-emulsion fluorescent type was found to be the most effective and is recommended for wrought products. Most of the tests were made on Al alloys.

Non-Destructive Testing. III. Use of Dye and Fluorescent Penetrants: Auxiliary Equipment for Local Examination. (*Aircraft Prod.*, 1958, **20**, Nov., 420–431). The examination of steel parts is included.

Electromagnetic Test Measures Effects of Shot Peening. P. M. Unterweiser. (*Iron Age*, 1958, **181**, June 5, 121–123). The use of a Russian-developed electrical induction type test for measuring the depth of a magnetic material affected by shot peening is described. It provides a quick and simple non-destructive method of checking the conditions of treated surfaces.—D. L. C. P.

How to Measure Magnetic Field Strength. L. Robjohns. (*SS News*, 1958, Sept., 16–18). An account of the Solus-Schall Precision Magnetic Field Meter.

On the Magnetic Properties of Quench-hardened Magnet Steel. T. Mishima, R. Hasiguti and Y. Kimura. (*Nippon Kinzoku*, 1958, **22**, Feb., 63–66). As the result of measurements of magnetic and mechanical properties of quench-hardened magnet steels, it was found that the coercive force did not change proportionally with Vickers hardness, but varied as the square root of the volume of austenite. Best magnetic properties were obtained when fine particles of martensite needles were distributed in the retained austenite, the ratio of the two phases being ∼ 2 : 1.

The Magnetization of Cobalt–Aluminium, Cobalt–Silicon, Iron–Aluminium and Iron–Silicon Alloys. D. Parsons, W. Sucksmith and J. E. Thompson. (*Phil. Mag.*, 1958, **3**, Oct., 1174–1184).

Defectoscope Examination of Rails in Moving Magnetic Fields. 2. Method of Modelling Electro-Magnetic Phenomena. V. V. Vlasov. (*Fiz. Met.*, 1058, **6**, (1), 74–81). The author discusses fundamental aspects of the theory of similarity and the similarity of magnetization of a rail by a moving field source is described.—R. S.

The Effect of the Order-Disorder Reaction on the Magnetic Anisotropy and Magnetostriction of Single Crystals of the Ferromagnetic Aluminium–Iron Alloys. R. C. Hall. (*Trans. Met. A.I.M.E.*, 1958, **212**, Oct., 703–706). The alloys examined were near the Fe_3Al composition. Variation in magnetostriction with ordering depended on composition; on disordering, anisotropy charged toward larger positive values for all compositions. Details of the experimental technique used are given.

Regarding the Arrangement of the Recrystallisation Structure in Cold Rolled Electrotechnical Steel. V. V. Druzhinin and I. P. Kudryavtsev. (*Stal', Supplt. Metallogr. and Heat Treatmt.*, 1958, 88–94). In specimens with a magnetic induction of more than 18.000 gauss most of the crystals are so arranged that the edges of the cubes are oriented along the direction of rolling with a variation of ± 5°. In specimens with a magnetic induction between 17 500–18 000 gauss the arrangement of the edges of the cubes takes place in the plane of the sheet, however, the majority of the crystals are oriented not in the direction of rolling, but at an angle of 10–15° to it. The reason for this difference in behaviour needs further investigation.—L. H.

Cube Oriented Magnetic Sheet. G. W. Wiener and K. Detert. (*J. Metals*, 1958, **10**, Aug., 507–508). A description is given of the nature, uses and magnetic properties of cube-oriented high-Si iron transformer sheet.

New Magnetic Steel. (*Met. Treatment*, 1958, **25**, Aug., 314). A note on Westinghouse "Cubex" oriented silicon–iron.

Single Crystal Iron Whiskers Adapted to the Basic Study of Magnetic Domains. G. G. Scott and R. V. Coleman. (*Gen. Motors Eng. J.*, 1958, **5**, July-Aug.-Sept., 34–37).

Surface Energy of a 90° Domain Wall in Iron. C. D. Graham, jun. (*J. Appl. Phys.*, 1958, **29**, Oct., 1451–1453). The calculation is used to predict the stable wall position in a whisker and agreement is reasonable.

Investigating Magnetisation Processes in Silicon Iron Crystals. Ya. S. Shur and V. R. Abel's. (*Fiz. Met.*, 1958, **6**, (3), 556–563). Under the influence of the magnetic field complicated changes arise in the structure of silicon–iron. These changes depend on the magnetic structure, on the strength of the field and on its orientation relative to the crystallographic axes. They may also occur with a change of boundaries, with a change of the direction of magnetization in some regions, with a change of dimensions etc. The pattern of these changes is of great importance for laminated magnetic material with an iron-type lattice and a sufficiently large constant of anisotropy.—L. H.

Effect of Elastic Stresses on the Magnetic Structure of Silicon Iron Crystals. Ya. S. Shur and V. A. Zaikova. (*Fiz. Met.*, 1958, **6**, (3), 545–555). The changes as a function of stress were investigated by powder patterns. The test material contained 3·5%Si. The findings show that the considerable changes in the magnetic structure of the crystals which may occur depend on the stress, and on its orientation with respect to the magnetic axes of the crystals. The changes may be irreversible. They can be satisfactorily explained by assuming that under the influence of elastic stresses, a redistribution takes place of parts in the crystal where the boundary energy has a minimum value and the tendency to easy magnetization nearest to the direction of the stress remains very pronounced.—L. H.

Temperature Dependence of Magnetic Properties of Silicon-Iron. C. W. Chen. (*J. Appl. Phys.*, 1958, **29**, Sept., 1337–1343). The range 30–700° C was covered for 3% singly oriented Si–Fe. An interpretation is offered.

Effect of Scaling on Magnetic Properties of Hot Rolled Steel for the Electrical Industry. V. V. Druzhnin and L. K. Kurennykh. (*Fiz. Met.*, 1958, **6**, (1), 40–45). In the hot rolled transformer steel, which was investigated, the drop in the specific gravity

of the steel caused by scale was 0·06 gauss/cc. Magnetic permeability of the scale in weak magnetic fields is similar to that of the steel while, in medium and strong fields, it is much lower. A drop in the field specific losses due to pickling with a 10 000 gs induction is negligible.—R. S.

Experimental Studies on the Formation of Superlattice in FeCo Alloy. III. Magnetic Properties. T. Yokoyama. (*Nippon Kinzoku*, 1956, **20**, Dec., 700–704). The coercive force increases as the temp. is reduced below 730° C, and increases suddenly at 550° C; the remanent magnetization shows similar behaviour. The curve of saturation magnetization shows breaks at 550° C and 730° C. The influence of field cooling is effective when cooling from 570° C (but not from 500° C) down to 200° C.

An Investigation on Boron-Treated Steels. VIII. IX. On the Effects of Boron-Addition to the Magnetic Properties of "Fe–Mn" Magnet-Steel. I, II. Y. Imai and H. Imai. (*Nippon Kinzoku*, 1956, **20**, Dec., 676–679; 679–682). VIII. The addition of as little as 0·003%B to magnet steel containing 13–15%Mn increases the max. energy product to 500 from 400 × 10^3 gauss-oersted, while the coercive force or residual induction is hardly affected. Boron added to the extent of 0·05% is undesirable, as it raises the austenitising temp. IX. Small amounts of boron added to a 12%Mn steel confer good magnetic properties. The coercive force and residual induction are increased considerably when the steel is rolled into thin sheets, as compared with blocks, rod or wire. Ni or Cr do not improve the properties of boron-treated Fe–Mn steel.—K. E. J.

Low-Temperature Transition of Magnetic Anisotropy in Nickel–Iron Ferrite. N. Menyuk and K. Dwight. (*Phys. Rev.*, 1958, **112**, Oct. 15, 397–405).

Magnetic Anistropy Constant of Yttrium Iron Garnet at 0° K. B. R. Cooper. (*Phys. Rev.*, 1958, **112**, Oct. 15, 395–396).

Electrical Conductivity of a Binary System of Vanadium Pentoxide with Ferric Oxide and Cupric Oxide and of a Number of Complex Alloys. O. A. Esin, V. L. Zyazev. (*Izvest. Akad. Nauk Otdel. Tekh.* 1958, (6), 7–11). The specific electrical conductivity in quasi-binary systems V_2O_5–Fe_2O_3 and V_2O_5–CuO was studied in the solid and liquid states. It was shown that all the alloys were semiconductors.—R. S.

Non-destructive Testing. (*Steel*, 1958, **142**, June 23, 110–112). The six basic methods of non-destructive testing (resonance, pulse reflection ultrasonics, eddy current instruments, radiography, penetrants and magnetic particles) and their fields of application are reviewed.—A. O.

Non-Destructive Testing. Part I. (*Aircraft Prod.*, 1958, **20**, Aug., 314–322). Ultrasonic inspection methods are described and the coding of defective material is shown. The various methods of using probes are illustrated and the savings obtained are estimated.

The Ultrasonic Transmitter-Receiver Probe. (*S. S. News*, 1958, Sept., 12–13). **Introduction to Ultrasonic Testing.** (13–15).

Applications of Ultrasonics in Metallurgy. J. Ors Martinez. (*Inst. Hierro Acero*, 1958, **11**, Jan.-Mar., 24–38). [In Spanish]. A review of the subject is given and particular reference is paid to the testing and control of materials and to the measurement of the elastic constants. (25 references).—F. S.

The Problem of Residual Stresses and Risk of Brittle Failure in Welded Constructions. H. Gerbeaux. (*Soud. Techn. Connexes*, 1958, **12**, March-April, 109–141). The theory of brittle failure is discussed and the effects of residual stresses on notches are referred to Residual welding stresses and stress relieving are dealt with. Tests for the control of brittle behaviour of steel are discussed. It is suggested that a service temperature-section thickness diagram for a given steel be used to determine when a stress relieving treatment is necessary. (48 references).—R. P.

Crack Propagation. W. MacFarlane. (*J. West. Scotland I.S.I. Paper* 518, 1959, Jan. 16, pp. 19). A review of previous work is correlated with measurements made with Stanton and Amsler-type test-pieces. The mechanism involved is discussed.

Investigations of the Low-Temperature Stability of the Steel SKhL-4 after Various Types of Heat Treatment. L. A. Gaivoronskii, E. V. Shustitskaya and K. V. Popov. (*Metall. Obra. Met.*, 1958, (9), 54–55). The steel contains 0·12%C, 0·77%Mn, 0·9%Si, 0·022%S, 0·024%P, 0·84%Cr and 0·56%Ni. The brittleness range was examined by impact testing and tendency to ageing was examined by 10% extension and 2 h tempering at 200° C. The serious effects of overheating are noted and the effects on welding pointed out.

A Method for Studying the Hydrogen Embrittlement of Steel. M. A. Figel'man and A. V. Shreider. (**Zavods. Lab.*, 1956, **22**, (5), 586–588). A device for precision bend testing of thin sheet specimens, the thinner the better, is described. Electrochemically loaded U9 steel strip was tested and brittleness was related to current density of polarization.

Studies on Heat-Checking of Iron and Steel. IV. Effects of Forging Ratio and Direction of Fiber Structure on Heat-Checking. T. Fujiwara. (*Nippon Kinzoku*, 1957, **21**, Oct. 573–576). The formation and propagation of heat-checks are associated with parts of fibre structure corresponding with boundaries of primary crystals. As forging ratio increases, the total number and depth of checks decrease, but the max. depth and mean depth hardly change; these effects are more marked parallel to the fibre direction. Specimens from the core of the billet are inferior as regards heat-checking. Homogenized and upset specimens have few directional properties, and are superior to normal forgings.—K. E. J.

The Comparison of Wear between One-Way Friction and Reciprocating Friction. I. Wear-Testing Machines and Their Characteristics. II. On the Wear of Steel Against Steel. S. Saito, K. Okabayashi and F. Nakamura. (*Nippon Kinzoku*, 1956, **20**, Nov., 600–604; 604–608). I. Machines designed to test one-way friction and reciprocating friction are described. II. With (carbon) steel against steel, wear characteristics depending on friction velocity and contact pressure are similar; as these parameters increase, the initial wear decreases and the state of stationary wear begins earlier. The amount of stationary wear with reciprocating friction is smaller than with one-way friction, but the difference is small. Hence, with steel, wear phenomena in reciprocating friction may be deduced from tests with one-way friction.—K. E. J.

Measuring Machine Tool Wear. E. I. Goodman. (*Atomic World*, 1959, **10**, Jan., 28–30). The use of pile-activated materials is described.

Effect of Heat Treatment on the Abrasive Wear Resistance of Steels for Earth-Moving Equipment. N. M. Serpik and M. M. Kantor. (*Metallov. Obra. Met.*, 1958, (7), 46–50). A method of testing rate of wear by soil is described and 7 grades of steel are tested. The heat-treatments of these steels are given. Hardness alone is not a guide to wear-resistance, and in all samples isothermal annealing was more effective than quenching and tempering to the same hardness. Steel U12 (1·15–1·24%C, 0·15–0·35%Mn, 0·15–0·35% Si, 0·2% max. Cr and Ni, 0·03% max. S), brought isothermally to acicular troostite with excess carbide and Vickers hardness 388–415 was the most resistant, 1½ times that of ploughshare steel heat-treated to Rockwell 58–62.

Some New Instruments for Continuous Testing of Materials. L. Beaujard and V. Husarek. (*Schweisstechn.*, 1958, **12**, April, 39–43). The authors discuss mechanization in sonic, magnetic and photoelectric methods of testing steel products.—R. P.

Review of the Various Methods of Non-Destructive Testing. A. de la Calle Estrada. (*Anal. Mec. Elect.*, 1958, **35**, Sept.-Oct., 246–252). Methods of non-destructive testing are discussed, their underlying principles, operation, advantages and limitations being dealt with.

Methods of Non-Destructive Testing. E. Mittli. (*Maschinenwelt Elek.*, 1958, **13**, July, 261–262). The paper provides a short summary of X-ray, γ-ray, ultrasonic and magnetic apparatus for non-destructive testing. Portable apparatus is referred to in particular. The advantages and limitations of each method are briefly discussed.—R. P.

Conference on Magnetism and Magnetic Materials. (*J. Appl. Phys.*, 1958, **29**, Mar.). **Time Decrease of Permeability in Iron.** G. W. Rathenau. (239–242). **Uniaxial Magnetic Anisotropy Induced in Fe–Ni Alloys by Magnetic Anneal.** E. T. Ferguson. (252–253). **Magnetic Resonance of Ferrites with a Compensation Temperature.** J. Pauleve. (259–263). **Some Results of an Electron Microscopical Study of the Metallographic Structure of Two Alloys for Permanent Magnets. Ticonal G and Ticonal X.** J. J. de Jong, J. M. G. Smeets and H. B. Haanstra. (297–298). **Relation between Colloid Pattern and Permanent Magnet Precipitate during the Magnetization Reversal in Alnico V.** K. J. Kronenberg and R. K. Tenzer. (299–301). **Phase Analysis of Alnico V Based on Temperature Effects.** R. K. Tenzer and K. J. Kronenberg (302–303). **Physical and Magnetic Properties of Elongated Single-Domain Iron and Iron–Cobalt Permanent Magnets.** R. C. Lever, E. J. Yamartino and R. B. Falk (304–306). **Effects of Magnetic Fields upon Anistropic Iron Crystals.** J. H. L. Watson, A. Arrott and M. W. Freeman (306–308). **Magnetic Measurements on Some Precipitating Systems.** A. E. Berkowitz and P. J. Flanders (314–316). The β-brass-iron system is included. **Magnetic Anistropy Induced by Magnetic Annealing and by Cold Working of Ni_3Fe Crystal.** S. Chikazumi (346–350). **Effect of Sample Thickness on the Field Annealing of of 6·5%Si–Fe.** P. A. Albert (351–352). **Effects of Composition and Processing Variables on the Magnetic Properties of the 50% Nickel–Iron Alloy.** M. J. Savitski (353–355). **Development of Preferred Orientations in Silicon Iron.** J. R. Brown (359–360). **(110)[001] Secondary Recrystallization Texture in Three Percent Silicon–Iron.** H. C. Fiedler (361–362). **Magnetic Properties of Cube Textured Silicon Iron Magnetic Sheet.** J. L. Walter, W. R. Hibbard, H. C. Fiedler, H. E. Grenoble, R. H. Pry and P. G. Frischmann (363–365). **Low Residual Induction in High-Aluminium Iron Alloys.** D. Pavlovic and K. Foster (368–369). **Effect of Elastic Bending on Magnetic Properties of Oriented Silicon Iron.** (370–371). **Effect of Hydrostatic Pressure and Temperature on the Magnetic Properties of a Nickel–Zinc Ferrite.** C. Q. Adams and C. M. Davis, jun. (372–373). **Preferential Volatilization of Cations from Ferrites During Sintering.** J. M. Brownlow. (373–375). **Some Properties of Quenched Magnesium Ferrites.** D. J. Epstein and B. Frackiewicz (376–377). **Ionic Valences in Manganese–Iron Spinels.** A. H. Eschenfelder (378–380). **Substitution for Iron in Ferrimagnetic Yttrium–Iron Garnet.** M. A. Gilleo and S. Geller (380–381). **Effect of Thermal History on the Antiferromagnetic Transition in Zinc Ferrite.** D. M. Grimes and E. F. Westrum, jun. (384–385). **Some Aspects of Tempering 3¼% Silicon Steel as Followed by Time Decay of Permeability.** E. S. Anolick and J. Singer (412–413). **Magnetocrystalline Anisotropy of Mg–Fe Ferrites: Temperature Dependence, Ionic Distribution Effects, and the Crystalline Field Model.** V. J. Folen and G. T. Rado (438–440). **Effect of Cation Vacancies on the Magnetic Annealing of Cobalt-Substituted Magnetite.** L. K. Bickford, jun., J. M. Brownlow, and R. F. Penoyer (441–442). **Magnetization Processes in Heat-Treated Single Crystal Cobalt Ferrite.** S. Foner and J. O. Artman (443–444). **Exchange Anisotropy in the Iron–Iron Oxide System.** W. H. Meikeljohn (454–455). **Domain Wall Motion in Metals.** R. W. DeBlois (459–467). Iron whiskers were used. **Effect of Cobalt on the Relaxation Frequency of Nickel–Zinc Ferrite.** F. J. Schnettler and F. R. Monforte (477–478). **Transitions from Ferromagnetism to Antiferromagnetism in Iron–Aluminium Alloys.** H. Sato and A. Arrott (515–517). **Further Magnetic and X-Ray Diffraction Studies on Iron-Rich Iron–Aluminium Alloys.** A. Taylor and T. M.

Jones (522–523). **Magnetic Domain Patterns on Iron Whiskers.** R. V. Coleman and G. G. Scott (526–527). **Domain Observations on Iron Whiskers.** R. W. DeBlois and C. D. Graham, jun. (528–529). **Antiferromagnetic Domain Walls and the Magnetization Process in α-Fe_2O_3.** I. S. Jacobs and C. P. Bean (537–538). **Size Dependence of the Wall Characteristics in a Two-Domain Iron Particle.** H. Amar (542–543).

Temperature Dependence of the Hysteresis Loops of An Iron–Chromium Alloy with 24%Cr. A. von Kienlin. (*Arch. Eisenh.*, 1958, **29**, Oct., 661–662). The form of the hysteresis loop obtained with an Fe–Cr alloy containing 24%Cr verified the result of previous susceptibility measurements made on a series of Fe–Cr alloys.

Magnetic Domains in α-Fe_2O_3. H. J. Williams, R. C. Sherwood and J. P. Remeika. (*J. Appl. Phys.*, 1958, **29**, Dec., 1772–1773).

On the Measurement of Permittivity and Tensor Permeability of Ferrites. N. Ogasawara. (*Tokyo Univ. Mem. Fac. Techn.*, 1958, (8), 513–523). [In English].

Permanent Magnets in Recent Years. N. Makino. (*Tetsu to Hagane*, 1958, **44**, May, 583–596). [In Japanese]. A review, with 59 references.—K. E. J.

Magnetic Properties of Spheroidal Graphite Cast Iron. Y. Shirakawa and T. Miyazaki. (*Nippon Kinzoku*, 1957, **21**, Jan., 22–24). At the same degree of graphitization, the maximum induction, coercive force and hysteresis loss of spheroidal graphite iron are lower, and max. permeability and residual induction are higher, than those of flake graphite iron. With increase in graphitization, all these parameters increase, except max. permeability which decreases.—K. E. J.

The Influence of the Method of Demagnetization on the Reversible Permeability of a High-Permeability Nickel Iron Alloy. R. C. Jackson, E. W. Lee and A. G. H. Troughton. (*Brit. J. Appl. Phys.*, 1958, **9**, Dec., 495–497).

An Apparatus for Magnetic Anisotropy Analysis. B. B. Timofeev. (**Elektrichestvo*, 1957, (5), 72–74). The distortion of the field produced by an alternating current by a rectilinear conductor connected at right-angles to the surface of the specimen is observed in the apparatus described.

Grain-oriented Electrical Steels. B. D. Jenkins. (*Elect. Times*, 1959, **135**, Jan., 3–7; A review of structure and properties.

Experimental Studies on the Formation of a Superlattice in FeCo Alloy. I. Equilibrium Specific Electric Resistance. II. Specific Heat and Hardness. T. Yokoyama. (*Nippon Kinzoku*, 1956, **20**, Nov., 644–647). I. From measurements of equilibrium specific electric resistance, it is found that the FeCo superlattice is quickly formed as soon as the temp. falls below the order-disorder transformation point. The curve of equilibrium of specific resistance against temp. is reversible, with breaks at 730° and 550° C. (13 references). II. For ordered FeCo alloy, the curve of sp. heat against temp. has peaks at 550° and 730° C. The transformation heat-temp. curve has a break at 550° C. With microhardness there is a max. at 525° C. The alloy thus has a change at 550° C.—K. E. J.

On the Interpretation of Errors in Sonic Testing. H. Krainer and E. Krainer. (*Schweisstechn.*, 1958, **12**, March, 28–32). The authors consider that reproducibility of results is within tolerable limits. The ratio of the echo due to imperfections to that due to normal reflection depends on the magnitude, shape and distance of the imperfection. The sensitivity of the test decreases linearly with section size.—E. P.

Failure of High-Carbon Steel Billets Detected by Ultrasonic Inspection. (*Tetsu to Hagane*, 1958, **44**, Mar., 347–349).—K. E. J.

On the Attenuation of Ultrasonic Waves in Steels by Residual Stress. I. Influence of Thermal Stress in a Cylindrical Steel Bar. II. Influence of Tensile Stress. A. Takaoki. (*Nippon Kinzoku*, 1957, **21**, Oct., 589–593; 593–597). I. Residual stress has a great influence on the attenuation of ultrasonic waves. Waves of 1–5 Mc/s were used on specimens of 0·38%C steel of size 70 mm dia. × 280 mm. With water-quenching from 700° C attenuation decreased at the centre and increased markedly at the circumference. A change produced by residual stress was found with water-quenching from 200–700° C, and another produced by stress-relieving appeared when quenched specimens were heated at 200–700° C. II. Waves of 3 Mc/s were used on 0·21–0·26%C and 25–20 stainless steels. Attenuation by tensile stress corresponds with the stress-strain diagram, but the effect is small. Tensile stress within the elastic limit causes a little attenuation in ferritic steel but has no effect with 25–20 steel. Attenuation in plastically deformed steels increases with tensile stress.—K. E. J.

On the Attenuation of Ultrasonic Waves in Ni–Mo–V Steel with Various Structures. A. Takaoki. (*Nippon Kinzoku*, 1957, **21**, Jan., 14–18). Results are given for 3 Mc/s waves in 3% Ni–Mo–V steel. Large austenitic grains cause the greatest attenuation, even even if the background structure is bainite. Transmission is weakened by a ferritic network. The pearlitic structure formed by isothermal transformation shows better transmissibility than coarse austenite in a bainitic structure.—K. E. J.

A New Method of Measuring Compressive Forces Using Ultrasonics. H. Krachter. (**Werkstatt und Betrieb*, 1958, **91**, (5), 246–248). Two parallel plates pressed together have a partial permeability to ultrasonics which is proportional to pressure. Apparatus making use of the effect is described and measurements in the ranges 3–30 and 200–2000 kg/cm² are used as examples.

Comparative Study of Defects in Heavy Forgings by Ultrasound and Betatron. L. Brand. (*Stahl Eisen*, 1959, **79**, Jan. 8, 32–36). The study confirmed that non-destructive testing of defects in heavy forgings that could impair their service is easily accomplished by both methods and that the results obtained agree. The study confirmed also the extreme sensitivity of the ultrasonic test and showed that it is necessary only in special cases to submit the workpiece to a betatron examination for supplementing the results.—T. G.

On the Attenuation of Ultrasonic Waves by Residual Stress (3rd Rep.) Influence of Compressive and Bending Stress. A. Takaoki. (*Nippon Kinzoku*, 1958, **22**, Feb., 81–84). Slight changes in attenuation of ultrasonic waves were observed in ferritic steels, but not in austenitic ones, both compressive and bending stresses causing the attenuation to decrease. **(4th Report); Attenuation by Thermal Stress in large Specimens and Austenitic Steel.** (84–87). Large specimens exhibited the same effects as small ones. The attenuation became less in the presence of thermal stress not only in ferritic but also in austenitic steels. The author concludes from this and from earlier work on the effect of residual stresses, that the mechanism of the change of attenuation must be different for the two types of stress.—J. G. W.

An Ultrasonic Irradiation Method of Immersion Testing of Metal Sheet and Comparison with Frequency-Modulation Methods. I. Egri. (*Acta Techn.*, 1958, **21**, (3–4), 327–345). [In German]. The "first-signal" method is described and shown to permit either detection of smaller flaws or a much greater speed of inspection. It can only be used for plates less than 8 mm thick and is limited by surface roughness.

Attenuation of Ultrasonic Waves in Metals. II. Stainless Steel. T. Hirone and K. Kamigaki. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1958, **10A**, Aug., 276–282). [In English]. A steel with 19%Cr and 10%Ni was used and increasing attenuation with frequency and grain size was observed. An explanation is advanced.

X-Ray Crystallography. J. Baxter. (*S. S. News*, 1958, Sept., 19–23). The use of a dekatron-gated counter unit is described.

The Reciprocal Links of Some Spectral X-Ray and Magnetic Characteristics of Alloys on an Iron Base. S. A. Nemnonov and K. M. Kolobova. (*Fiz. Met.*, 1958, **6**, (3), 466–474). The alloy elements were: Armco iron (99·8%), Al (99·96%), and chemically pure zinc. Fe–Al alloy with 10; 18; 25: 50 and 75%Al was prepared in a high-frequency vacuum furnace. Fe–Zn alloy was prepared with a thin iron foil 7–12 micron thick, weighed and introduced in a quartz ampoule with the correct quantity of zinc so that on vaporization and diffusion an iron–zinc alloy of required composition should result. The ampoule was unsealed *in vacuo* and kept for 30–35 min at 620–680°. The strong asymmetry of the Ka_1 and Ka_2 spectral lines of the atoms of the migrating elements and the corresponding magnetic characteristics are conditioned chiefly by the number of unpaired 3*d* electrons. The numerical values of the asymmetry index for the Ka_1 line can be utilized as an index characterizing the magnetic moment of the atoms of the migrating metal.—L. H.

A Simplified Procedure for Calculating Peak Position in X-Ray Residual Stress Measurements on Hardened Steel. D. P. Koistinea and R. E. Marburger. (*A.S.M. preprint* 102, 1958, pp. 14). The method described is based on theoretical factors, which affect diffraction peak symmetry. Correlations are found between this method and that of Christenson and Rowland and also with a mechanical dissection technique.

Microradiography, Principles and Applications in Iron and Steel Research. W. Johnson and K. W. Andrews. (*Iron Steel*, 1958, **31**, Sept., 437–444). The basic principles and experimental technique of microradiography are described together with its relation to other modes of investigation. Some examples of its application are given.—A. G.

Experimental Studies on the Formation of Superlattice in FeCo Alloy. IV. Thermo-electromotive Force, Thermal Expansion and X-ray Investigation—I: V. On X-ray Investigation—2. T. Yokoyama. (*Tetsu to Hagane*, 1957, **21**, May, 321–324; 325–329). [In Japanese]. IV. If an ordered specimen is heated from room temp., the thermal e.m.f. decreases suddenly at 550° C and gradually from 550° to 730° C. The coeff. of thermal expansion shows a small increase at 550° C and a max. at 730° C. Fe–Co alloy has a body-centred cubic lattice in all ranges of superlattice formation. Changes in lattice constant correspond with those in expansion. V. Measurements with a Geiger-counter X-ray diffractometer show that order commences at 730° C and the degree of order increases suddenly at 550° C.—K. E. J.

Subgrain Structure in an Fe–Si Crystal as Seen by X-Ray Extinction Contrast. J. B. Newkirk. (*J. Appl. Phys.*, 1958, **29**, 995–998).

Gamma Radiography of Fillet Welds. S. C. Hall. (*S. S. News*, 1958, Sept., 7–8). Use of a ^{192}Ir source is described. **Supporting Gear for X-Ray Equipment.** (9–11).

The Effect of Gamma Irradiation on the Potential Behaviour of Platinum and Stainless Steel Electrodes. W. E. Clark. (*J. Electrochem. Soc.*, 1958, **105**, Aug., 483–485).

Neutron Diffraction Investigation of the 119° K Transition in Magnetite. W. C. Hamilton. (*Phys. Rev.*, 1958, **110**, June 1, 1050–1057).

How Radiation Damages Metals. M. J. Makin. (*Atomics Nucl. En.*, 1958, **9**, Sept., 318–322). Effects on creep rate, hardening of single crystals, stress-strain curve and on ductile-brittle transition are discussed.

The Application of the Discoveries of Nuclear Physics to the Metallurgical Industry of the U.S.S.R. (*United Nations, E/ECE/305; E/ECE/Steel/119*; 1958, 22–47). Uses of tracer elements for the study of diffusion, solid solution structures, distribution of elements in alloys, blast furnace and steel furnace processes, behaviour of sulphur and slag reactions, O.H. operation, and γ-radiography and rolling thickness control are briefly described. (89 refs.).

Determination of Equicohesive Temperatures. R. Tamhankar. (*Trans. Indian Inst. Met.*, 1958, **11**, Sept., 13–36). The comparative behaviours of coarse and fine grain materials and stress concentration effects of intergranular fissures formed above the equi-

cohesive temperature were studied for an austenitic Ni–Cr steel. For Armco iron most of the methods did not apply on account of secondary effects. Structures are illustrated and the significance of the results is discussed.

Susceptibility to Volatilisation of a FeNi Alloy in the Temperature Range 77 to 290° K. N. Z. Miryasov. (*Fiz. Met.*, 1958, **6**, (1), 188–190). Determination of the relationship between temperature and the volatilization process of a FeNi alloy in the range 77° to 290° K showed that volatilization is inversely proportional to $H^{1/2}$ (H being the value of the magnetic field in Oersted).—R. S.

Symposium on Elevated Temperature Strain Gages. (*A.S.T.M.S.T.P. No.* 230, 1958, June, pp. 163). Uses for aeronautical purposes were covered primarily but the properties of all types are discussed as well as methods of application and uses.

An Apparatus for Heating Small Samples to Elevated Temperatures at High Pressures. J. Fennell, N. H. Hancock and R. S. Barnes. (*Met. Treatment*, 1958, **25**, Aug. 332–334). A device incorporating a pressure vessel capable of withstanding 6000 psi at 950° C is described. A furnace is lowered over the end containing the sample for heating purposes. The internal diameter is $\frac{5}{8}$ in. The apparatus was designed for the study of the effects of temperature and pressure on gas bubbles and voids in solids.

Differential Thermal Analysis Above 1200° C T. F. Newkirk. (*J. Am. Ceram. Soc.*, 1958, **41**, Oct., 409–414). Equipment and techniques for use up to 1550° C are described. The $CaO–Al_2O_3–Fe_2O_3$ system is used as an example and the analysis of heating curves is discussed.

Industrial Applications of X-Ray Analysis. (*Proceedings of 6th Annual Conference* 1957, *Denver, Colorado*, pp. 494). **The New Hilger Microfocus X-Ray Generator and its Applications.** F. Brech and J. R. Stansfield (17–36). **A Simplified Procedure for Calculating Peak Position in X-Ray Residual Stress Measurements on Hardened Steel.** D. P. Koistinen, R. E. Marburger and R. P. Mitchell (37) (Abstract only). **Amorphous Solids, Small Particles and Thin Surface Films.** F. Schossberger (73–99). Examples include the study of electroless Ni coatings and the nature of the black stain produced on cold-rolled annealed steel. From these and other results a scheme for the reactions involved is elaborated. The film consists of oxides and free C. **Electrolytic Isolation and the X-Ray Study of Micro-constituents in Steels.** K. W. Andrews and H. Hughes. (101–127). Isolation is described and the cells and electrolytes used are shown and choice of apparatus and medium is indicated. After isolation X-ray studies of composition, particle size and shape, and lattice parameters are explained and interpretations are discussed with examples. Tempered martensite relating carbide to stage of tempering, carbides of low-alloy steels, sigma-phase, carbide reactions in Cr–Mo steel and other cases are briefly considered. **A Theoretical Derivation of the Composition of Mixable Specimens from Fluorescent X-Ray Intensities.** J. Sherman (231–251). The example of $ZnO–Fe_2O_3–TiO_2–Al_2O_3$ is considered. **X-Ray Spectrographic Determination of Tantalum, Niobium, Iron and Titanium Oxide Mixtures Using Simple Arithmetic Corrections for Interelement Effects.** B. J. Mitchell (253–269). **Determination of Uranium Dioxide in Stainless Steels by the X-Ray Fluorescence-Solution Method.** L. Silverman, W. Huk and L. Moudy (271–281). A rapid direct method after solution in $HClO_4$ is described. Sr is used as internal indicator. **Results Obtained with the Modified Norelco Autrometer.** D. C. Miller (283–295). Apparatus requiring only insertion of a sample and a standard giving the analysis of a stainless steel in ~ 2 min. is described. Magnesium and any elements of higher atomic n. can be determined. **Rock Analysis by X-Ray Fluorescence Spectroscopy.** A. A. Chodos, J. J. R. Branco and C. G. Engel (315–327). **Micro-Analysis by X-Ray Absorption, Fluorescence, Emission and Diffraction Using Ultra-Fine X-Ray Sources.** V. E. Cosslett, P. Duncumb, J. V. P. Long and W. C. Nixon (329–337). A brief account of the Cambridge work. Studies on steel are not referred to here. **Applications of the Electron Probe Microanalyzer.** L. S. Birks and E. J. Brooks (339–349). Uses on Cu–Fe minerals and in mass transfer studies in liquid metal cooling systems are described. **X-Ray Spectrographic Analysis of Nickel Containing Alloys with Varied Sample Forms.** C. M. Davis and G. R. Clark (351–366). The examination of drillings, millings and filings is described. **Analysis of Diffusion Couples by X-Ray Absorption.** R. E. Ogilvie (439–453). The process is described and examples are given, one of which is the Fe–Cr couple. **Components for X-Ray Polychromators.** J. L. Jones, K. W. Paschen, H. H. Swain and G. Andermann (471–480). Use of the device can be made to determine S in steel with a deviation of $\pm$ 0·005%. **Recent Applications of the X-Ray Microscope.** S. E. Summers (483–494). Ductile iron with nodular graphite is one example shown.

On the Problem of Standardising Radiographic Apparatus. G. Moravia. (*Schweisstechn.*, 1958, **12**, March, 32–37). The author discusses the desirability of standardising radiographic apparatus and the principles involved. Photographic and dosimetric measurements are compared. The latter are indirect but offer the advantages of being more quantitative and less empirical.—R. P.

Micro-Radiography by Means of Photo-Electrons from Excited Metallic Surfaces. J. M. Bermudez de Castro y Mosquera. (*Inst. Hierro Acero*, 1958, **11**, July-Sept., 180–184). [In Spanish]. A brief account of the theory and practice of this technique is given and illustrated by two examples, one of cast iron and the other of a particle thrown out of a basic convertor during the blow.—P. S.

A Micro-Analyser with Electronic Probe. C. Crussard. (*Schweisstechn.*, 1958, **12**, April, 43–47). The author discusses the applications of the electronic micro-analyser with particular reference to applications in the ferrous field. Localized analysis makes it possible to determine inclusions, sulphides, carbides and oxides, light elements excepted, and to show up segregation. The analyser can be used as an aid to the determination of equilibrium diagrams. The error is about 1% and the lower limit of detection 0·2–0·3%.—R. P.

Radio-Spectrographic Control of the Structural Condition of Lifting Chains. A. Homès and R. Pankowski. (*Rev. Mét.*, 1958, **55**, Sept., 865–871). A radio-spectrographic control test described by the authors shows that the characteristic brittle fracture often developed in the curved portion of the links of wrought iron or steel, occurs in the regions of greatest work-hardening. This work-hardening and subsequent ageing causes the brittle fracture.

Radiographic Investigation of Strains and Particle Sizes in Iron and Steel. I. Influence of an Elastic Stress in Iron and Comparative Measurements on Tungsten. F. Brasso and H. Möller. (*Arch. Eisenh.*, 1958, **29**, Dec., 757–771). Intensity distributions in X-ray interference patterns were measured on a plain carbon steel after various heat treatments, in order to determine the lattice condition of polycrystalline Fe in respect of distortions and particle sizes. A goniometer and scintillation counter were used for the intensity measurements; iron samples in the stressed and unstressed conditions were tested on samples given a stress-removing anneal at 650° C, Mo K_α radiation being used. Tests were also made on a specimen annealed at 800° C, and comparative measurements were made on a sintered W specimen. The results are discussed.

The Structure of the Energy Spectrum of Electrons in Iron–Chromium Alloys. N. D. Borisov, V. V. Nemoshkalenko and A. M. Fefer. (*Doklady A.N.*, 1958, **121**, (2), 264–267). The aim of this work was to study the significant changes in the X-ray spectra, especially in the region of the K-group, of the Cr and Fe atoms on their transition into melt and their transformation from α- to σ-phase. The main results are: (1) the energy of the bottom of the band of conductivity of pure Cr is increasing with temperature; in iron, there is a break at 850° C in the direction of higher energy; (2) the energy of the ceiling of this band of pure Cr and Fe remains constant at 600°–1200° C; (3) Fermi energy of Fe–Cr melts does not depend on the concentration of the components. Numerical data are presented in a table and a diagram.—S. I. T.

The Use of Radioactive Isotopes in Metallographic Research. K. Sagel. (*Metall*, 1959, **13**, Jan., 25–30). A survey of the possible uses of radioisotopes in metallographic research is given, with reference to the literature. A table of isotopes is included, with half-life values and limits of sensitivity. (116 refs.).

Report on Elevated-Temperature Properties of Chromium Steels. (12–27 per cent). W. F. Simmons and H. C. Cross. (*A.S.T.M. S.T.P. No.* 228, 1958, July, pp. 113). Reports are presented on 12Cr, 12Cr–2Ni, 13Cr–2Ni–3W (Greek Ascoloy), 12Cr–3W, 12Cr–Nb, 12Cr–2·75Mo–V (Lapelloy), 12Cr–2·5W–V, 12Cr–5Co–3W–V (Cobalt Ascoloy), 13–Cr–W–Mo–V (422), 12Cr–2W–2Mo–V (422M), 12Cr–Mo–Nb–V (H–46), 14Cr, 16Cr–2Ni, 17Cr, 18Cr, 27Cr, 27Cr–Mo and 6 other 12 Cr steels.

High Temperature Properties of Four Wrought Steels Used in the Chemical Industry. B. J. Connolly and G. Boyd. (*Chartered Mech. Eng.*, 1958, **5**, June, 263–264). A summary is given of creep and stress-rupture tests on low-C 1%Cr–Mo steel, medium-C 1%Cr–Mo steel, low-C 3%Cr–Mo steel and austenitic Cr–Ni steel with 2·5%Mo and 0·5% Ti.

The Deformation and Fracture of Alpha–Iron at Low Temperatures. W. D. Biggs and P. L. Pratt. (*Acta Met.*, 1958, **6**, Nov., 694–703). Stresses for slip, twinning and cleavage of single crystals of α–Fe were determined at − 183° C. It is shown that brittle fracture in the case of α–Fe crystals is described by the dislocation theory of cleavage, twin boundaries acting as the necessary dislocation boundaries. Twinning results from a burst of dislocations released at the upper yield point and is considered to precede fracture; elimination of twinning by pre-straining or decarburization eliminates fracture by cleavage.

Some Low Temperature Observations of 1020 Steel. F. S. Deronja and M. Gensamer. (*A.S.M. preprint* 87, 1958, pp. 11). It was found that an increase in the grain size lowered the ductile-to-brittle transition temperature and that the latter was raised considerably when the testing speed was increased. Twinning preceded crack nucleation and propagation in the brittle range. Twinning was the primary mode of deformation below the transition temperature in coarse-grained steel, but slip was more important in finer grained metal.—E. E. W.

The Mechanism of Diffusion in the Solid State. C. E. Birchenall. (*Met. Rev.*, 1958, **3**, (11), 235–277). Fick's law, diffusion mechanisms in metals, the Kirkendall effect, temperature dependence, activation energy and the frequency factor are considered and then ionic crystals, oxides with NaCl structure, spinel oxides, covalent crystals, and boundary and surface diffusion are reviewed. Self-diffusion coefficients are tabulated for metals, S and P. (192 refs.).

Volumes of Activation for Diffusion in Solids. R. W. Keyes. (*J. Chem. Phys.*, 1958, **29**, Sept., 467–475). Effects of pressure on diffusion are reviewed and activation volumes for diffusion are calculated using simple models of lattice defects. A general equation which represents experimental findings with fair accuracy can be derived from several of the models in terms of activation energy and compressibility and Zener's expression can be obtained from it.

Some Observations Regarding the Present Status of Measurements of the Diffusion Coefficients of Hydrogen in Iron and Mild Steel. R. C. Frank. (*J. Appl. Phys.*, 1958, **29**, Aug., 1262–1263). The observation of two sets of diffusion coefficients for H_2 in iron, one after the gas had been in the metal

for some time, is discussed. The respective activation energies are 3000 and 6000 cal/mole.

The Reduction of Passive Films by Hydrogen Diffusion through Steel. R. T. Davis jun. and T. J. Butler. (*J. Electrochem. Soc.*, 1958, **105**, Oct., 563–568). A technique for following diffusion of H_2 by electrode-potential measurements is described. Specimens of tin-plate iron strip were exposed on one side to H_2SO_4 and on the other to $Na_2Cr_2O_7$ solution and changes of potential on the chromate side caused by penetration by H_2 were measured. The changes (of from 300–900 mv) were believed to arise from reduction of the passive film. Changes of thickness, composition and solution concentration were correlated.

Kinetics of Hydrogen Absorption Desorption and Permeation of Metals. T. G. O. Berg. (*Corrosion*, 1958, **14**, Dec., 562t–566t). Absorption from moist H_2 and from dilute acids and the kinetics of absorption, desorption and permeation processes are given. Migration and diffusion are considered and compared with measured values.

Hydrogen Behaviour of Sheet Steel: A Report of Experimental Work Using Glass Coatings. G. P. K. Chu. (*Sheet Metal Ind.*, 1958, **35**, Aug., 585–588, 619). The effect of H_2 on steel is reviewed and results of experiments on the chipping of glass coatings on steel due to H_2 diffusion and precipitation are reported. A cell for H_2 generation and diffusion is described and illustrated. The relation of diffusion of atomic hydrogen through a metal lattice to the susceptibility to corrosion is pointed out.—T. G.

Activity of Oxygen in Liquid Iron Alloys. T. P. Floridis and J. Chipman. (*Trans. Met. A.I.M.E.*, 1958, **212**, Aug., 549–553). The activity coefficient of dissolved O_2 is established as $\log f_0 = -0{\cdot}20\,[\%O]$. Additions of Co, Mo, W and Pt increase the coefficient, and Cu and Au decrease it. (19 ref.).

The Effect of Vanadium on the High and Low Temperature Mechanical Properties of a 1 Cr–1 Mo Cast Steel. L. D. Tote and R. S. Zeno. (*Mod. Castings*, 1958, **34**, Aug., 87–94). Step additions were made to 10 000 lb heats raising the C content from 0·04 to 0·29% in 1–Cr–1–Mo and from 0·23 to 0·54% in 1–Cr–1–Mo–¼V steels and the products were tested in normalized and tempered conditions at comparable hardness levels. Increased V lowered rupture ductility up to 0·22%, but it remained constant with larger amounts. Tensile and yield strengths were unaffected. Bainite became finer to 0·4% and coarser with 0·54%, grain size increased up to 0·4% and decreased again with 0·54%. Fe_3C and Mo_2C were replaced by finer V_4C_3. Creep strength appears to fall off with 0·54% V.

Mechanical Properties of Fe–Al–Si Alloys at Room and Elevated Temperatures. D. J. Schmatz and V. F. Zackay. (*A.S.M. preprint*, 1958, (111), pp. 11). An increase in Si at all Al levels was found to increase the strength, the electrical resistivity and the stress-rupture properties, but to decrease the ductility. Si appeared to be more nearly additive than Al in improving the oxidation resistance and also in initiating order. A correlation was found to exist between ordering temperature and creep rate.—E. E. W.

A Brief Survey of the Influence of Alloying Elements in Stainless Steels. K. W. Bee. (*Murex Rev.*, 1958, **1**, (19), 545–570). The steels are classified and the effects of Cr, Ni, Mn, Si, Mo, W, Ti, Nb, Al, Cu, Co and V and S and Se are discussed in detail with numerous diagrams of constitution and physical properties. A section on intergranular corrosion is added with compositions and treatments for its prevention.

Rare Earths Homogenize Ingots. W. G. Wilson. (*Iron Age*, 1958, **182**, Oct. 2, 78–79). Reference is made to recent Russian and American work on the effects of rare earths on iron and steel. Improved ductility, hot workability and homogeneity are reported and possible explanations are discussed.—A. G.

The Nature of the Interaction Between Aluminium Nitrogen, Boron, and Titanium as it Influences Some Properties of Steel. L. L. Pyatakova and Ya. E. Gol'dshtein. (*Fiz. Met.*, 1958, **6**, (2), 347–352). Investigation into the deformability, breaking strength, microstructure and difference in grains were examined for the influence of these elements in the solid phase. The simultaneous effect of Al, B, and N depends to a large extent on the order of their introduction in the steel. In molten steel deoxidized with Al, boron does not form boron nitrides because the nitrogen in steel is combined with Al forming aluminium nitrides. This ensures the presence of boron in the solid solution and hence its influence on the qualities of steel. Upon introduction of, or saturation with nitrogen, the boron-containing steel partially or totally eliminates boron from the solid solution owing to the preferential formation of the carbon nitrides. The small tendency of boron steels to overheating may be due in the first place to an inhibitory action of the titanium carbides on the growth of austenite grains.—L. B.

Effect of Boron on the Structure and Properties of Steel. M. V. Pridantsev and O. N. Meshcherinova. (*Stal', Supplement Metallogr. and Heat Treatmt.* 1958, 3–20). Boron lowers the critical points for steel both in heating and in cooling and in the proportion of up to 0·01% reduces also the temperature of the intensive growth of the austenite grains. For good ductility the steel in the bath should be deoxidized with Al before the addition of ferro-boron and 0·06% titanium to combine with nitrogen. An increase in the hardening temperature beyond the optimum reduced ductility, but a high homogenizing temperature and prolonged heating for forging does not.

Investigation of the Effects of Incongruous Elements and the Interaction Effects of These Elements on the High Temperature Strength of Fe–Co–Ni–Cr Alloys. J. H. Sye, T. L. Robertshaw and F. M. Richmond. (*WADC Techn. Report*, 57–426; *ASTIA Document*, AD142237; *PB* 131614 250, 1957, Dec., pp. 102). Vacuum melted alloy with 50 atomic %Ni, 20%Cr, 10%Fe and 10%Co was examined with five combinations of other elements. Al, Ti and W additions to cast or forged alloy gave stress-rupture lives of 133·9h at 1800° F and 149h at 1750° F respectively at 15 000 psi. Metallography is described.

Metal Selector. (*Steel*, 1958, **143**, Oct. 20, 81–16). A tabulated list is presented giving the compositions, forms available, properties and typical use of leaded steels, vacuum melted metals, spring alloys, cast high alloys, copper alloys, high temperature alloys, high-strength and superstrength steels.—A. G.

Materials Selector: Irons and Steels. (*Mat. Design Eng.*, 1958–59, *Reference Issue*, **48**, Mid. Oct., 33–65). Tabulated properties of nodular or ductile irons, malleable irons, ingot and wrought irons; low, medium and high carbon steels, free-cutting steels, Mn, C–Mo, Ni–Cr–Mo, Cr–V, Ni–Cr–Mo, Mn–Si and Mn–Ni–Cr–Mo steels; Cr–Mn stainless, austenitic, martensitic and ferritic stainless steels; age-hardening stainless steels; cast carbon steels, cast low-alloy steels, cast stainless steels, cast heat-resistant alloys; nitriding steels and A.S.T.M. pressure vessel steels are given.

Alloy Steel Plates. (*A.I.S.I. Steel. Prod. Man.*, 1958, Sept., pp. 38).

How to Get More For Your Special Alloy Dollar. (*Iron Age*, 1958, **182**, Oct. 16, 185–200). A review is made of 8 major types of special ferrous alloys, namely spring steels, silicon steels for cores, magnetic core materials, cast alloys for heat resistance, corrosion resistance castings, austenitic manganese steels, special steels for nitriding and the 200 series of stainless steels.—A. G.

Steels, Alloy and Carbon. E. W. Husemann. (*Met. Prog.*, 1958, **76**, Oct., 120–123). A brief review of the newer alloy steels with a table of the A.I.S.I. standard compositions. Leaded steels and nitriding and cold forming, high strength and super strength grades are also referred to. The article is part of a review of everyday uses of metals.

Cast Age-Hardenable Austenitic Steels. E. A. Lange, N. C. Howells and A. Bukowski. (*Mod. Castings*, 1958, **34**, Oct., 53–59). Use of Cr–Ni–P, Cr–Mn and Cr–Ni–Mn–V steels were studied as materials for high-strength non-magnetic castings. An alloy of good ductility was developed with a yield strength of 100 000 psi. A Mn–V type was also developed with a minimum of alloying elements.

Study of Austenitic Chrome Manganese Steels. A. S. Panicheva. (*Metallov. Obra. Met.*, 1958, (5), 23–27). The steels were tested for use as fasteners for high-pressure steam pipes. Compositions, heat treatments and structures of 18 experimental and 5 semi-industrial materials are given. One with 14%Cr and 18%Mn had good creep and relaxation properties. After 1000 h at 550–650° C its impact value was steady at 4½ kgm/cm². Addition of 1·2%W, 0·17–0·7%Ti and 1·5%Nb impairs the properties by promoting σ-phase formation, which decomposes to sigma. During ageing of a Cr–Mn–V steel precipitation at boundaries of austenite occurs with increase of strength.

Special Alloys Boost Metalworking's Future. (*Steel*, 1958, **143**, Oct. 20, 158–164). A review is given of the properties and applications of various types of special alloys including vacuum-melted metals, cast high alloys, high temperature alloys and superstrength steels.

Consider Cr–Mo–V Low Alloy Steels for 1000 F Service. C. M. Cosman. (*Mat. Design Eng.*, 1958, **48**, Oct., 93–96). Mechanical properties of carbon steels, Cr–Mo, Cr–V, and Cr–Mo–V steels are tabulated.

High-Temperature Alloys: 1900–1958. F. S. Badger. (*J. Metals*, 1958, **10**, Aug., 512–516). The development and processing of high-temperature alloys in relation to the development of the gas turbine is reviewed in a reproduction of three papers presented by the A.I.M.E. High-Temperature Alloys Committee.

Precipitation-Hardening Steels Up to 1200° F. F. K. Lampson. (*S.A.E. Trans.*, 1958, **66**, 565–573). Properties of 17–7PH, 17–4PH, AM350, AM355 and A–286 steels are given. The heat treatment of each grade is described and notes on welding and brazing are added.

Designing with ph Steels. (*Prod. Eng.*, 1958, **29**, Sept., B8–B11; from SAE paper). The composition, mechanical properties, heat treatment and welding and brazing properties of precipitation-hardening steels are discussed.

Which High Strength Steel? R. J. Nekervis, C. H. Lund and A. M. Hall. (*Mat. Design Eng.*, 1958, **48**, July, 84–89). A review of the properties of hot-work tool steels, martensitic stainless, martensitic low-alloy hardenable, precipitation-hardenable semi-austenitic stainless steels. A note on the cold-rolled austenitic stainless type is added. Advantages and disadvantages are listed and developments noted.

Ultra-Strength Steels. (*Prod. Eng.*, 1958, **29**, Sept., 1313 from *Climax Molybdenum Bulletin*). A description of the properties and treatments of the ultra-strength steels.

Application of Low-Alloy High-Tensile Steels. I. M. Mackenzie. (*Iron Coal Trades Rev.*, 1958, **177**, Oct. 10, 859–863). Reasons for the small demand for low-alloy steels for construction are discussed. The conditions in which they might be suitable are summarized after a discussion of fatigue, brittle fracture, corrosion and fabrication problems.

Ultra High Tensile Steel. L. Marshall and H. Beresford. (*Engineering*, 1958, **186**, Oct. 3, 446–448). Studies on Rex 539 and its comparison with other steels, including U.S. materials are reported. Physical properties are given. Heat treatment may give rise to decarburization, and the use of protection by copper plating is described, though this may produce absorption of hydrogen. Forgings and tubes were also tested.

Super Strength Steel Hardens in Air. (*Steel*, 1958, **143**, Sept. 8, 104). The development of a new high strength sheet steel alloy by the U.S. Steel Corporation is reported. The alloy, Airsteel X-200 is air hardenable, developing a tensile strength of 280 000 p.s.i.

Stainless Steels. R. E. Paret. (*Met. Prog.*, 1958, **74**, Oct., 116–117). Part of a review

of everyday uses for metals. Various types, their extrusion and flow turning, the slag washing process, and sandwich-rolled and decorative products are reviewed.

Recent Research Into Stainless Steels. (*Mech. World*, 1958, **138**, Sept., 405–406). Notes on sigma-phase, the search for a 12%Cr ferritic age- or precipitation-hardening steel, conservation of Ni, stress-corrosion and embrittlement, and Fe–Cr–C–N and additions to 12%Cr steels.

Stainless Alloys—Present and Future. W. O. Binder. (*Chem. Eng. Prog.*, 1958, **54**, Nov., 45–50). A review of composition and corrosion resistance properties of old and new stainless steels and other alloys.

METALLOGRAPHY

Oxidation and Figures of Thermal Etching of Fe–Si Alloys in a Hydrogen Atmosphere Containing Traces of Water Vapour. A. Ferro, C. Sari and G. Venturello. (*Met. Ital.*, 1958, **50**, Nov., 484–494). The authors study oxidation as a fraction of time and temperature of Fe–Si alloys (1%, 3·5%, 6·4%) in H_2 containing traces of water vapour. Attention is focused on the formation of pure silica and in a more oxidizing atmosphere, to the formation of fayalite. At temperatures above 1200° C, the authors study the formation of etching figures on the sample surfaces. These figures can be related to the orientation of the base crystal. Attention is drawn to the oxidation of zones re-crystallized while maintained at high temperature. (26 References).

Investigations on Silica Inclusions in Cast Iron. G. Nañdori. (*Koh. Lapok*, Öntöde, 1958, **9**, Oct.-Nov. 225–235). The coagulation and rising of silica inclusions in liquid cast iron is promoted by the Mn content. Thus, silica inclusions present in a cast iron containing less than 1%Mn are of submicroscopic size, and they are of microscopic size only in a cast iron with 1%Mn or more. The Mn takes part, therefore, not only in desulphurization, but also in the deoixdation of cast iron.

Examination of Inclusions in Steel. J. Verö. (*Koh. Lapok*, 1958, **13**, Oct.-Nov., 449–459). The author describes the various inclusions in steel, their origin, and their visual (macroscopic and microscopic) and chemical examination. The following chemical examination methods are discussed: determination of oxygen in the form of Al_2O_3, vacuum hot-extraction, isolation of oxide, carbide and sulphide inclusions. Some adequate deoxidation methods for decreasing the oxide inclusions in steel are also given.

Studies on the Origin of Non-Metallic Inclusions (Preliminary Report). T. Tokuda. (*Tetsu to Hagane*, 1958, **44**, Mar., 321–323). X-Ray data show evidence for spinel, α-alumina and quartz in inclusions.—K. E. J.

Study on Non-Metallic Inclusions in Steel. VIII. On the Sand Mark in Steel. H. Takada. (*Tetsu Hagane*, 1958, **44**, Mar., 324–326).

A Study on Carbides in High Speed Steels. I. S. Kimura. (*Tetsu to Hagane*, 1958, **44**, Mar., 412–414). Chemical compositions of the steels are given, and also the composition of the carbide and the ferrite after various heat-treatments. Relationships are shown between tempering temp. and the concentration of metallic elements in the carbide.—K. E. J.

Electrolytic Potential in the Electrolytic Isolation of Carbides of Ball-Bearing Steels. M. Ueno and H. Nakashima. (*Tetsu to Hagane*, 1958, **44**, Apr., 483–486). When carbides are isolated, the matrix does not always dissolve ("peeling"). Quenching from higher temp. increases this effect—the retained austenite gives a nobler matrix. The potential is baser for annealed than for quenched steel. Peeling is strongly correlated with Cu content.—K. E. J.

A Study of the Carbides in Iron and Steel by Electrolytic Isolation. VI. On the Heterogeneities of Cementite in Low Alloy Steels. T. Sato and T. Nishizawa. (*Nippon Kinzoku*, 1957, **21**, Jan., 47–50). Results from steels containing 0·61%C, 1·08%Mn and 1·13%C, 1·42%Cr, show that elements which raise the A_1 point (Cr, Mo, W, V, Ti) dissolve more readily in pro-eutectoid than in pearlitic cementite, while those which lower it (Mn, Ni) dissolve less readily. (10 references). **VII. Carbides in Tempered Plain Carbon and Low Alloy Steels.** (June, 418–422). In the first stage of tempering, hexagonal close-packed carbide was pptd. In the third stage, orthorhombic carbide was pptd.; reflections resembled those from platelet cementite, and there was no evidence of iron percarbide. Small amounts of alloying elements made no changes. The nucleation of orthorhombic carbide is retarded by Si, Al and P, and hastened by S, Cr and V.—K. E. J.

Study on Carbides in Commercial Special Steels by Electrolytic Isolation. V. On Carbides in Several Cold-Working Die Steels. T. Sato, T. Nishizawa and K. Murai. (*Tetsu to Hagane*, 1958, **44**, May, 565–570). In annealed high-C high-Cr die steels the carbides are M_7C_3, while modified steels (3%W or 1%Mo) have mainly $M_{23}C_6$. High-C high-W steels contain $M_{23}C_6$ with some M_6C. Most of the Cr, W and Mo content of the steel is in the carbides; after normal quenching the C, Cr, W and Mo contents of austenite are less than those of the steel as a whole. During tempering these carbides are formed from ε-carbide via θ-carbide. **VIII. Carbides in High Speed Steels—II.** T. Nishizawa. (Mar., 414–416). Results are given for eight steels ranging from 18–4–1 to 5–9–1 steel. Hardness figures are given, and relationships are shown between the chemical composition of the carbides and the proportions of M_6C, MC and $M_{23}C_6$ in them.—K. E. J.

The Segregation of Epsilon-Carbide in Ferrite and Martensite During Tempering. W. Pitsch and A. Schrader. (*Arch. Eisenhütt.*, 1958, **29**, Nov., 715–721). After tempering massive steel specimens, containing supersaturated carbon in solution, particles were found to have been segregated in the ferrite consisting of two groups of parallel bands at right angles to one another. Electron diffraction examination showed that these were ε-carbide particles.—L. J. L.

Investigation of Oxide Inclusions Deformable by Hot Rolling in Mild Rimming Steels. H. Lessing and H. Malissa. (*Arch. Eisenh.*, 1958, **29**, Oct., 627–629; discussion 629). The needle-shaped oxide inclusions occurring in rolled mild rimming steel were isolated and examined by microchemical and radiographic methods. The melting point of the needles was 900–1000° C, consistent with their deformation at rolling temperatures. Their structure and composition are discussed.

Contribution to the Isolation of Sulphides in Steels. W. Koch and E. Artner. (*Arch. Eisenh.*, 1958, **29**, Dec., 737–744). It was shown in experiments on a series of commercial steels with S contents ranging from 0·005–0·27% that sulphides of Mn, Fe, Ti and Zr can be isolated quantitatively from an electrolyte with the composition: 5%Na citrate, 1·2%KBr and 0·6%KI. The effect of pH was investigated. Various methods of separation of the isolates from carbides etc. were examined, of which only magnetic separation was satisfactory. The structures of the isolated sulphides was studied.

Metallographic Examination of a Fusionless Weld on a Pearlitic Base. A. S. Amat. (*Cien. Tecn. Solda.*, 1958, **8**, Sept.-Oct., 1 (153). Comparative mechanical and metallographic examination of a "fusionless" and normal welding of specimens of pearlitic cast iron showed that the former type of weld was superior in strength and structure.

A Unique Solidification Structure from an Iron Melt. H. Schenck, H. Baumann and M. Nacken. (*Arch. Eisenh.*, 1958, **29**, Dec., 773–774). An unusual solidification structure was obtained from a melt under vacuum, and developed from a 48 g Fe nodule. The probable causes of the growth of the structure are discussed.

Studies on Duplex-Grain Structure of Austenite. II. Causes of Formation of Duplex-Grain Structures in Steels. Y. Masuko. (*Tetsu to Hagane*, 1958, **44**, Apr., 476–479; May, 559–564). "Uniform" duplex-grain structures appear during abrupt coarsening of austenite as the result of solution of grain-growth inhibitors. "Non-uniform" structures are caused by primary segregation of grain-growth inhibitors. Such inhibtors are primarily AlN, and to prevent duplex grains the content and primary segregation of this must be controlled. Other conditions which produce these structures are described. (24 refs.).—K. E. J.

Ferrite Indicator for Austenite Stainless Steel. T. Kamimura. (*Tetsu to Hagane*, 1958, **44**, Mar., 346–347).—K. E. J.

Studies on the Primary Structure of Steel. II. Primary Structure and Segregation. U. Masuko. (*Tetsu to Hagane*, 1958, **44**, Mar., 382–384). The effects of varying additions of Al and Al + Ti on forged and cast structures and degree of segregation of various steels (including 1·7%C and 0·4%C–Cr–Mo steel) are described.—K. E. J.

Relation between Austenitic Grain Size and Aluminium Nitride. IV. Effect of Hot Working and Thermal History on Austenitic Grain Size. G. Kato. (*Tetsu to Hagane*, 1958, **44**, Mar., 428–430). Results are given for S15 CK and SCM 21 steels; the grain size is followed from the ingot through heating, forging, rolling and various heat treatments.—K. E. J.

The Effect of Nitrogen on the Tendency to Growth of the Austenitic Grain Size of Medium-Carbon Phosphorus-Containing Steel. V. N. Svechnikov and I. Kh. Trush. (*Metall. Obra. Met.*, 1958, (9), 15–19). Steels with 0·035–0·089%P were used some deoxidized with Al as Al combines with N. They were melted in a h.f. furnace and saturated with N_2, the P content being adjusted with ferrophosphorus. The N content was raised as high as 0·025%. Grain size was correlated and the separation of aluminium nitride in the deoxidized steels was proved.

Formation and Stability of ε-Phase on Nitrided Iron. A. Burdese. (*Met. Ital.*, 1958, **50**, Dec., 529–537). The paper discusses morphological characteristics, mechanical properties and formation velocity of nitrided layers on different irons. The nitrogen distribution in the surface layers and the effects of quenching and tempering on the stability of the nitrides is examined with particular reference to the ε-phase. (15 References).—M. D. J. R.

Comparative Study of the Allotropic Transformations of Iron and Uranium. P. Lehr and J. P. Langeron. (*Rev. Mét.*, 1958, **55**, Sept., 829–839). The α–β and α–γ changes in Fe and U respectively both take place by a process of nucleation and growth; stresses produced by volume changes give rise in each case to true plastic deformation. The results obtained for each metal are compared.

On the Theory of the Irreversible Process in α-Iron. J. Meixner and B. Manz. (*Forsch. Wirts Nordrhein-Westfalen*), 1958, (492), ppl 10). Using the C atoms in α–Fe as an example, the irreversible process in α–Fe is considered, with a discussion of two specias cases.

Transformations During Heating of Hardened White Iron. F. K. Tkachenko and V. F. Zubarev. (*Metall. Obra. Met.*, 1958, (9), 24–26). Thermal, dilatometric and magnetic curves were obtained simultaneously during the heating and cooling of hardened and non-hardened iron. The apparatus is described, the curves being recorded photographically. Steels U10 and 60 S2 were also included. Magnetic transformation of the ferrite was observed below the Ac_1 point and at the second transformation temperature austenite gave martensite and a magnetic transformation of some kind took place in the carbide. Permeability increased in the third transformation range, supposed to be due to stress relief in the α-phase.

Influence of Various Elements on the Formation of Spheroidal Graphite in Cast Iron Treated with Magnesium. R. Ozaki. (*Tetsu to Hagane*, 1958, **44**, Mar., 359–361). The elements discussed are Si, Cr, P, Mn, Cu, Mo, Ti, Sn, As, Al and Te.—K. E. J.

Studies on Graphitization of White Cast Iron. II. The Effect of Copper. K. Futakawa. (*Nippon Kinzoku*, 1957, **21**, June, 382–385). Results are given for 2·35%C, 1·75%Si iron. With increase of copper

content in the range 0·32 → 1·15%, the time for completion of the first stage of graphitization was reduced, that for the second stage (direct transformation) was increased, and the temp. range in which graphitization was completed only by direct transformation was increased.—K. E. J.

The Free Energy of Formation and the Interfacial Enthalpy in Pearlite. J. J. Kramer, G. M. Pound and R. F. Mehl. (*Acta Met.*, 1958, **6**, Dec., 763–771). In the case of the Fe–C eutectoid, the interfacial enthalpy of the pearlite phase was found to be 1400 ± 300 ergs/cm², and the interfacial free energy estimated to be 700 ergs/cm², this being probably about ½ that required by the Zener theory for pearlite growth.

The Effect of Rate of Cooling from the Austenite Range upon the Microstructure of Low-Carbon Rimmed Steel. On the Formation of Pearlite-Free Structure. K. Yoshida. (*Tetsu to Hagane*, 1958, **44**, Apr., 471–475). Experiments were made with a range of cooling rates after homogenizing annealing at 950° C for 30 min. Pearlite-free structures were found in specimens cooled below 700° C at 3° C/min; these were of 0·08%C steel cooled at above 150° C/min and 0·05%C steel cooled at above 20° C/min to 700° C. These specimens showed very high Charpy impact values.—K. E. J.

Sub-zero Treatment of Quenched Steel. III. Some Considerations on the Stabilization of the Retained Austenite. Y. Imai and M. Izumiyama. (*Nippon Kinzoku*, 1957, **21**, July, 437–440). Results are given for several types of steel (analyses given). A suggested mechanism of stabilization of retained austenite after partial transformation to martensite is based on the increase in resistance to martensite formation attributable to interstitial solute atoms (especially C or N) forming clusters in the austenite. (38 references).—K. E. J.

Change in Some of the Properties of Steel EI 612 as a Function of the Extent of Decomposition of the Austenite. V. I. Prosvirin and L. F. Chernov. (*Metall. Obra. Met.*, 1958, (6), 10–14). Steel with 0·1%C, 14·3% Cr, 35·8%Ni, 2·6%W, 1·4%Ti and 0·37%Al was heated at temperatures of 650–1000° C for 1 h and hardness and impact-resistance were measured, also changes over longer periods at 750° C were recorded. Embrittlement was most intensive during the first 1–2 h. At high temperatures impact strength increases again on prolonged heating and is more rapid at higher temperatures. The phases separating and the changes in plastic properties are also described.

Oscillations and Resonance in the Kinetics of Martensite Formation. N. V. Grum-Grzhimailo. (*Doklady A.N.*, 1958, **121**, (5), 850–851). There is a tendency for the iron carbide in austenite to form a polymer $Fe_{18}C_6$, which is unstable on its separation from austenite and forms graphite. The velocity of the reaction $Fe_{18}C_6 \rightleftarrows 6Fe_3C$ of the isothermal breakdown of austenite can be calculated by means of a formula in which $\phi(T)$ is a temperature function relating to the changes in the outer electron shells of iron atoms followed by a rearrangement of electron spins and emergence of ferromagnetism. The curve for $\phi(T)$ can be obtained from the diagram of decomposition of austenite by an empirical setting of one of its factors. The curve obtained shows a marked oscillatory character and a strong resonance at the point of martensite formation.—S. I. T.

The Effect of Tensile Stress on the Austenite–Martensite Conversion in Iron–Nickel Monocrystals. A. Kochendörfer and G. Otto. (*Arch. Eisenhütt.*, 1958, **29**, Nov., 709–714). Large, uniformly oriented martensite areas were produced in Fe–Ni monocrystals with 30% Ni by slow cooling to the temperature of liquid air while maintaining a temperature gradient of 10–20° over the length of the sample under the action of tensile strain amounting to 75% of the yield point.—L. J. L.

The Transformations of Austenite: Mechanism of Pearlite and Bainite Formation. M. Doi and K. Fujimori. (*Bol. A.B.M.*, 1958, **14**, Jan., 111–138). [In Portuguese]. Theories are based on the Fe–C diagram theoretically calculated by one of the authors and on metallographic and crystallographic observations. Pearlite is formed by incoherent co-precipitation of the two phases, the ferrite first formed is supersaturated in carbon, but immediately decomposes to stable ferrite and cementite. Upper bainite is formed by (*a*) initial precipitation of cementite followed by (*b*) transformation of the surrounding regions into supersaturated ferrite by a coherent, diffusionless process. This supersaturated ferrite immediately transforms into stable ferrite and cementite with relief of internal stresses. Slip to relieve internal stresses cannot occur. Lower bainite forms by (*a*) initial precipitation of cementite (*b*) transformation of the surrounding regions into supersaturated ferrite by a diffusionless process, similar to martensite formation, with relief of internal stresses by slip, and (*c*) growth by plate formation similar to the martensite in In–Tl. Cementite precipitates preferentially in the slip planes of the ferrite from the supersaturated solution. The differences between upper and lower bainite formation are attributed to the atomic mobility of carbon atoms. (33 references).—P. S.

Sub-zero Treatment of Quenched Steel (4th Report). Control of Stabilisation of Retained Austenite. Y. Imai and M. Izumiyama (*Nippon Kinzoku*, 1958, **22**, Jan., 7–11). Austenite stabilization after partial transformation to martensite was investigated on 1%C steels containing respectively 0·14%Al, 0·04%Ti, 1%Ni, and 0·08%Sn. From the results it is concluded that stabilization is based mainly on the effect of interstitial solute atoms.—J. G. W.

Concerning the Formation of Austenite in Steel During Heating by the Reverse Martensite Transformation. B. K. Sokolov and V. D. Sadovskii. (*Fiz. Met.*, 1958, **6**, (3), 568–569). A metallographic proof is offered regarding the two possible mechanisms of austenite formation in quenched structural steel alloys. A diffusion mechanism similar to that occurring during the decay of the super-cooled austenite (pearlite-troostite decay) and the other, without diffusion through the martensite transformation when the austenite can originate either by a non-diffusion interaction between ferrite and carbide or by a non-diffusion process similar to the reverse martensite transformation.—L. H.

Martensitic Transformations and Determination of the Inhomogeneous Deformation. D. S. Lieberman. (*Acta Met.*, 1958, **6**, Oct., 680–693). A graphical method is presented for determining directly from observations of the crystallographic features of a martensitic transformation, the inhomogeneous deformation which forms part of the total transformation, and the pattern of which has seldom been observed. Evidence of the correctness of the method is given, and the technique is applied to a case in which accurate experimental data are available.

Effect of Preliminary Plastic Deformation on the Martensite Transformation in a FeCrNi Alloy. G. V. Kurdyumov, O. P. Kaksimova, A. I. Nikonorova, Z. D. Pavlenko and A. M. Yampol'skii. (*Fiz. Met.*, 1958, **6**, (1), 95–105). An account is given of investigations to explain the influence of plastic deformation and subsequent annealing on the martensite transformation in FeCrNi alloys of the 18/8 type. Varying degrees of deformation were applied at 20, 100 and 175° C. Plastic deformation can accelerate or retard martensite transformation depending on conditions of deformation and annealing. Slight deformation accelerates transformation, widens the temperature range in which transformation occurs, and increases the initial rate of isothermal transformation and the total amount of the martensite phase. When the amount and temperature of deformation are increased, the retarding effect of deformation on the austenite becomes more apparent.—R. S.

Cooling Transformation Diagrams for A.I.S.I. 4024 and 52100. R. C. Hess and D. J. Blickwede. (*Met. Prog.*, 1958, **74**, July, 96B). (*Metal Progress* Data Sheet).—T. G.

Transformation Diagrams for the Heat-Treatment of Steels. S. N. A. Narayan and G. Mayer. (*Tisco*, 1958, **5**, July, 120–128). The relationships between the nature of transformation products obtained in isothermal reactions and those derived by continuous cooling are discussed. Techniques obtaining continuous cooling diagrams are described with examples, and the application of the curves to heat-treatment problems is outlined.

Phase Changes in Steel in Relation to the Iron–Carbon System. J. S. Vatchagandhy. (*Tisco*, 1958, **5**, July, 85–91). A general account of the phase changes in the Fe–C system at different temperatures and the normal heat-treatment procedures employed for steel.

Transformations in Steel during Slow Cooling. P. K. Chakravarty. (*Tisco*, 1958, **5**, July, 92–105). An account is given of the structures obtained by the slow cooling of steel, the mechanism of their formation and the mechanical properties obtained (40 refs.).

Phase Transformations in Steel during Rapid Cooling. V. G. Paranjpe. (*Tisco*, 1958, **5**, July, 106–119). The mechanisms of the changes taking place in the rapid cooling of steel from the austenitizing temperature are discussed, and the requirements for rapid cooling for hardening deduced. (53 refs.)

Influence of the Non-Uniform Electrical Resistance of Steel on Phase Transformations with Electro-Thermal Treatment. N. M. Rodigin. (*Fiz. Met.*, 1958, **6**, (2), 315–320). The special features of the electro-thermal treatment are the non-uniformity of the precipitation of substances in the micro-volumes which have a different specific electrical resistance. The spread of the thermal energy depends on the form and on the distribution of these microzones. The deformation of the crystal lattice is therefore a result of the unequal heating and it also contributes to austenization.—L. H.

Metallurgical Problems. Equilibrium Diagrams. J. A. Morales Belda. (*Dyna*, 1958, **33**, Feb., 75–93). [In Spanish]. The use of binary equilibrium diagrams to determine solidification temperatures, microstructures etc. is explained by means of a series of questions and answers related to a variety of alloy systems.—P. S.

The Redistribution of a Second Phase during Annealing in a Temperature Gradient. P. G. Shewman. (*Trans. Met. A.I.M.E.*, 1958, **212**, Oct., 642–647). The redistribution of phases in a 2-phase system due to the existence during annealing of a temperature gradient is discussed, and the possible mechanism examined. Fe_3C is shown to be transported to the hot end of α-Fe, and the kinetics of the distribution are calculated, making certain assumptions.

Liquid Immiscibility in Copper–Iron and Copper–Cobalt Systems in the Supercooled State. Y. Nakagawa. (*Acta Met.*, 1958, **6**, Nov., 704–711). The constitution of Cu–Fe systems was studied by magnetic measurements at high temperatures and by microscopical examination of the quenched specimen. Liquid immiscibility was found only in the supercooled state; the temperature of critical mixing was found to be 20° C below the liquidus curves at the equi-atomic composition.

CORROSION

The Present Status of the Oil Ash Corrosion Problem. NACE Task Group T-5B-3. (*Corrosion*, 1958, **14**, Aug., 369t–372t). Corrosion has been attributed to V_2O_5, S or NaCl. Five case histories are discussed and proposed solutions are offered using inhibitors, different alloys, lower temperatures or changed oil composition.

Corrosion Phenomena in Condenser Tubes. A. J. Maurin. (*Corros. Anticorros.*, 1957, **5**, Dec., 383–393; 1958, **6**, Jan., 15–25). The author summarizes the various reasons which have been advanced for the corrosion of condenser tubes, particularly those of cupronickel and of iron. He goes on to discuss the action of $KMnO_4$ as an inhibitor and the theory of cathodic protection. Preventive measures as well as cures are suggested.—R. P.

Corrosion of Iron Alloys by Superheated Steam. T. Samuel. (*Corros. Anticorros.*,

1958, **6**, Jan., 3–8). The author reviews work carried out on the high-temperature oxidation of iron alloys and the theories on which calculations on the rate of oxidation are based. The influence of alloying elements and time are taken into account. This is followed by a description of Cebedeau's thermogravimetric method for studying the corrosion of some alloys in superheated steam. (20 references).—R. P.

Some Relations between Deposition and Corrosion Contamination in Low Make-Up Systems for Steam Power Plants. E. S. Johnson and H. Kehmna. (*Corrosion*, 1958, **14**, Sept., 429t–434t). A mechanism is advanced to suggest that the ratios and interactions of contaminants are more significant with regard to deposition than the absolute amounts of the individual contaminants. Examples are quoted. The ratio of Cu to Fe seems to be significant, small variations in other ions do not seem to be important.

Removal of Corrosion Products from High-Temperature, High Purity Water Systems with an Axial Bed Filter. R. E. Larson and S. L. Williams. (*Corrosion*, 1958, **14**, Sept., 424t–428t). The removal of radioactive corrosion products by an axial bed filter developed for the purpose is described.

Corrosion of Steel Weldments. J. U. MacEwan and H. H. Yates. (*Corrosion*, 1959, **15**, Jan., 2t–6t). A test on 46 steel weldments by periodical immersion in 3%NaCl for 12 months is reported. Weight lost depended on type of steel rather than type of weld, but in the weld region both factors were concerned. Values were obtained for the short circuit current between surfaces of parent and weld metal. Oxygen was bubbled into the solution and the activation energy increased from 4 to 6 kcal mole with fall of O_2 content in the gas mixture from 100 to 20%.

A Review of the Gas Condensate Well Corrosion Problem. D. R. Fincher. (*Corrosion*, 1958, **14**, Sept., 403t–404t). A review with special reference to the high pressure well is given. The various methods of applying inhibitors are noted and the relationships to well completion are related to the method of protection used.

Corrosion of Valve Seat Ring Results in Flexible Metal. J. E. Prior. (*Corrosion*, 1959, **15**, Jan., 1t). The failure of a 316 stainless steel ring is described. It is attributed to neglect of homogenizing heat treatment.

Observations on Corrosion Resistance of High Strength Stainless Steels for Aircraft. J. Halbig and O. B. Ellis. (*Corrosion*, 1958, **14**, Aug., 389t–395t). Comparison of precipitation-hardening types is made including a new stainless alloy PH 15–7 Mo. Tests on high-temperature high-purity water resistance, effects of nitriding, atmospheric corrosion resistance, resistance to sea water, galvanic corrosion, stress cracking, and brazing and welding are reported.

On the Influence of Aggressive Carbonic Acid on the Corrosion of Steel in Water. E. Uusitalo and J. Heinänen. (*Tekn. Kemian*, 1958, **15**, Dec. 31, 625–627). [In English]. Tests on specimens of steel pipe are reported showing the effect of the hardness of water on corrosion in presence of CO_2. The effect of soft water is shown to be much more considerable than of hard. Up to a limiting concentration the amount of CO_2 in hard water is shown to be non-corrosive.

Corrosion of Mild Steel in Aqueous Monoethanolamine. H. R. Froning and J. H. Jones. (*Ind. Eng. Chem.*, 1958, **50**, Dec., 1737–1738). Tests with pure solutions at b.p. in atmospheres of N_2 and CO_2 were made in presence of H_2S. H_2 was evolved and silica (from glass vessels) was found to reduce corrosion.

Weight Loss of Common Cast Iron by Corrosion with Fused Alkali. K. Imada. (*Tetsu to Hagane*, 1958, **44**, Mar., 365–366). Details are given of weight loss against soaking time in NaOH at 500° C. The weight loss decreases with repeated immersions.—K. E. J.

Corrosion Problems in the Manufacture of Phosphoric Acid from Elemental Phosphorus. J. C. Barber. (*Corrosion*, 1958, **14**, Aug., 357t–362t). Corrosion rates of various metals are measured using 316 stainless steel as standard. The standard was examined over a range of concentrations, temperatures and locations and compared with type 201 and type 430 17–4 PH steels and non-ferrous alloys. There appear to be locations where the more expensive steels would be economic.

Corrosion and Water Purity Control for the Army Package Power Reactor. R. J. Clark and A. L. Medin. (*Corrosion*, 1958, **14**, Sept., 419t–423t). The water for a pressurized system is demineralized to a resistivity > 600 000 ohm/cm. Corrosion problems are reviewed, the rate has been lower than the estimate of 0·05 mg/cm²/month.

The Distribution of Soil Conductivities and Some Consequences. G. N. Scott. (*Corrosion*, 1958, **14**, Aug., 396t–400t). Soil resistivity data are given and without advanced mathematical techniques it is shown that there is a normal distribution of the logs of soil conductivity over a large range of soil types. Practical applications of the relationship are discussed.

Exfoliation of Metals at High Temperatures. K. Inagaki. (*Sci. Rep. Res. Inst. Tôhoku Univ.*, 1958, **10A**, Oct., 388–399). [In English]. A number of metals were tested. In the case of steel, a forged sample was cut, sliced, crushed and put into a screw pulverizer, which can be heated up to 500° C. The grain size of the powder produced depends on the temperature. The experiments and the results are described in detail.—T. G.

Growth in Needles of the Hematite Phase in Films Formed in the Oxidation of Iron at Elevated Temperatures. J. Paidassi. (*Acta Met.*, 1958, **6**, Dec., 778–780). [In French]. Details are given of experiments on the production of whiskers in the Fe_2O_3 phase of films formed on Fe heated to various temperatures. The conditions favouring the production of the whiskers were established.

Methods of Corrosion Testing. W. Wiederholt. (*Corros. Anticorros.*, 1958, **6**, March, 80–100). Tests carried out in moist air, in SO_2 and CO_2 at various temperatures, in saline vapours and by the method of alternate immersion are described. The results of these tests are used to find the method which will give results comparable with service conditions.—R. P.

Results of Co-ordinated Experiments by Several Laboratories for the Development of Corrosion Testing Methods Capable of Standardisation (I). W. Wiederholt. (*Metalloberfläche*, 1958, **12**, Oct., 305–307; Nov., 337–347; Dec., 369–371). The object of the experiments is outlined and the testing conditions described in detail. The materials tested, method of carrying out the tests and experimental results from individual laboratories are dealt with. Finally the experimental results are discussed.

Corrosion by Condensed Moisture—A Simple Test Apparatus. S. C. Britton. (*Corros. et Anticorros.*, 1958, **6**, Oct., 346–351). The apparatus described exposes plate samples to conditions corresponding to covered storage in a humid atmosphere. The resistance of various metals and metallic coatings to this type of corrosion is discussed.—A. G.

A Method for Determining Corrosion Rates from Linear Polarization Data. M. Stern. (*Corrosion*, 1958, **14**, Sept., 440t–444t). The slope of the linear portion of the Fe polarization curve can be related to corrosion rate. A theoretical basis for this observation and the testing method using this relationship is now worked out and its uses are discussed.

Electrochemical Measurements to Determine the Causes of Damage Due to Corrosion. H. Schmecken. (*Werks. Korr.*, 1958, **9**, Nov., 690–693). Three cases of corrosion are studied electrochemically under operating conditions, by examining C.D. curves plotted on log-linear paper.

An Electrochemical Study of Pitting Corrosion of Stainless Steels. N. D. Greene and M. E. Fontana. (*Corrosion*, 1959, **15**, Jan., 32t–38t). An artificial pit specimen has been devised and growth of the fault in 18%Cr–8%Ni steel characterized. The autocatalytic nature of the process is confirmed and evidence has been obtained of ion screening at pit sites for the first time. **2. Polarization Measurements.** (39t–44t). Pitting with the above specimen in $FeCl_3$ solution was examined and found to be a complex, time-dependent process. A new concept based on a mixed potential theory is advanced.

Electrochemical Study of the Corrosion of Stainless Steels. J.-M. Defranoux. (*Corros. et Anticorros.*, 1958, **6**, Oct. 363–372). The bases and application of electrochemical methods of corrosion testing are discussed, with special reference to the localized corrosion of stainless steel.—A. G.

Intensiostatic and Potentiostatic Investigations of the Electrochemical Behaviour and Corrosion of Metals and Alloys. (*Rev. Soc. Roy. Belge Ing. et Industr.*, 1958, Nov., 499–500). Report of a Study Group held in July at the University of Brussels.

Corrosion Testing Using Aerosols. W. Hess. (*Tidsskr. Kjenic Bergv. Met.*, 1958, **18**, (10), 161–165). The salt spray test is criticized and a new apparatus using a finely dispersed homogeneous liquid medium is described and evaluated. It appears to reproduce natural conditions much more nearly.

The Acetic Acid Salt Spray Test. J. H. Hooper. (*Bull. Inst. Met. Fin.*, 1958, **8**, Spring, 79–90). A comparison of a 1% acetic–5% salt spray test with other accelerated methods of testing on Cr–Ni–Cu systems on steel and non-ferrous systems is made, and outside exposure tests were also carried out. The 24h SO_2 test was found most suitable for steel, but the acetic acid test served better for non-ferrous metals. Tests with 5% salt alone did not correlate with rate and type of breakdown in outside exposure.

A Radioactive Corrosion Probe. D. Anson. (*J. Inst. Fuel*, 1959, **32**, Jan., 10–14). A device for use in air heaters and boiler economizers and other heat-exchange systems is described. It is internally cooled and the corrosion is estimated from the activity of the washings after test.

Effect of Grain Size on the Resistance of Steels to Cavitation Erosion. I. N. Bogachev and R. I. Mints. (*Metallov. Obra. Met.*, 1958, Aug., 26–29). Resistance to cavitation erosion is shown to depend on grain size, the nature of the grain boundary and the nature of the grain itself. Further work is needed.

Corrosion Rates in Port Hueneme Harbor. C. V. Brouillette. (*Corrosion*, 1958, **14**, Aug., 352t–356t). A steel and a stainless steel were included in the tests and an immersion exposure for 30 months was made. The mode and rates of corrosion are given and are typical for California sea water.

Corrosion of Metals in Tropical Environments. C. R. Southwell, B. W. Forgeson and A. L. Alexander. (*Corrosion*, 1958, **14**, Sept., 435t–439t). Tests over 8 years in the Canal Zone are reported and compared with exposures of the same materials in temperate zones. The effects of low-alloy additions are recorded. Copper steel and 2% and 5%Ni steels and four proprietary low-alloy steels are reported on. The copper steel was only 13–14% less corroded than mild steel, the Ni steels were very effective.

Contribution to the Evaluation of the Qualities of Cast Irons Corrosion-Resistant at Various Temperatures. P. Detrez. (*Corros. Anticorros.*, 1957, **5**, Dec., 374–382). The author stresses the importance of stating the conditions under which a part is to operate in order to choose a suitable material. Low-P cast irons are referred to in particular. The relation between the temperature and application are tabulated against structure and composition. Carbon content of about 3·4% acts as a stabilizer against thermal shock. Mn of the order of 1% stabilizes pearlite while high P gives a tendency to graphitization.—R. P.

Study of the Intercrystalline Corrosion of 18/8 Stainless Steel by Observation of Internal Friction. M. Smialowski and W. Drozd. (*Corros. Anticorros.*, 1957, **5**, Dec., 394–397). The authors describe their experiences with Kê's apparatus, using copper sulphate and also boiling nitric acid as the corrosive media.

They conclude that the determination of changes in internal friction offers no advantage for studies on intercrystalline corrosion. However, the method offers possibilities for studying those structural changes in austenitic steel which render it more prone to intercrystalline corrosion.—R. P.

Tests on the Corrosion of Electrolytically Deposited Zinc coatings on Chemically Passivated Steel. T. Biestek. (*Corros. Anticorros.*, 1958, **6**, Feb., 59–65). The author discusses tests and results regarding the relative merits of 15 passivating solutions for zinc coatings. Chemical passivation increases the corrosion-resistance of zinc coatings. Among solutions recommended for passivation is one containing chromic, nitric, sulphuric and acetic acids.—R. P.

Study of Internal Corrosion at Varying Pressures and the Distribution of Purified Town Gas. J. Morlet. (*Corros. Anticorros.*, 1958, **6**, April, 127–131). Tests and results are given regarding the corrosion of mild steel by purified town gas at various pressures. It was found that dehydration of the gas and/or reducing its oxygen content will render it less corrosive.—R. P.

The Effect of Aeration on the Corrosion Resistance of Welded Joints in 18Cr–9Ni Steel in Boiling Sulphuric Acid. N. A. Langer and B. I. Medovar. (*Avtom. Svarka*, 1958, **11**, (11), 48–51). Experimental data are given which evidence the property of O_2 to delay the corrosion of welded joints in 1Kh18N9T (18Cr–9Ni) steel in boiling H_2SO_4. The delaying action was also promoted by small additions of HNO_3.—M. T.

On the Passivity of Stainless Steels in Sulphuric Acid. Influence of Thermal Treatment between 1050 and 650° C and Influence of Cold Deformation. C. Carius. (*Mét. Corr. Ind.*, 1958, **33**, Jan., 6–29). The author describes results and tests on 18/8 type stainless steels; the potential during corrosion is taken as a measure of the free energy change during dissolution in $4NH_2SO_4$. The description of the method of potential measurement is followed by a discussion on the influence of pH and of thermal treatment on the transformation potential. Pitting due to chloride ions is described, also the electrochemical behaviour of steel in calcium chloride after cold deformation.—R. P.

Documentation and Rust Protection by Paints. K. A. van Oeteren-Panhäuser. (*Werks. Korr.*, 1958, **9**, Nov., 681–686). Methods of recording and evaluating laboratory and field corrosion tests are discussed.

On the Theory of Iron Oxidation. B. Prenosil and J. Polasek. (*Hutn. Listy*, 1958, **13**, (10), 908–916). After a review of the present state of knowledge of the process of iron oxidation, the authors outline their concept of the relationships between partial O_2 pressures on the grain boundaries and the concentrations of crystal lattice defects. The work is primarily devoted to an analysis of the kinetics of the growth of polyphase oxide layers. The growth of individual layers is not necessarily independent of that of other layers, the relationship between which can be quantitatively determined by the method of analysis indicated. The process of the formation of the oxide layers during an isothermal process is governed by the parabolic law. Through analysis the diffusion coefficients of the migratory components of the oxide layers and the differences in their concentration on the grain boundaries may be determined.—M. T.

An Investigation of Intergranular Oxidation in Stainless Steels and High-Nickel Alloys. I. C. A. Siebert, M. J. Sinnott, L. H. DeSmyter and R. E. Keith. (*WADC Tech. Report* 55–470; *PB* 131471; 1955, June, pp. 100). Type 310 stainless was oxidized for 100-h periods under stress and up to 500-h unstressed and intergranular corrosion was observed microscopically. The scales formed were examined with X-rays and the subsurface by electron diffraction. The products in surface and subsurface regions were found to be substantially identical.

Further Studies On the Metallurgy of Silicon Iron: Some Observations on Selective Oxidation. A. V. Seybolt and E. I. Alessandrini. (*Trans. Met. A.I.M.E.*, 1958, **212**, Aug., 507–509). An investigation to identify the oxide films formed on iron containing 3½%Si at low O_2 potentials, and to examine the H_2O/H_2 ratios required to form them is reported.

Method of Determination of Strength of Oxide Films. V. N. Gulyaev. (*Zavods. Lab.*, 1958, (10), 1245–46). The method of formation of oxide films on the flat surface of contact of two cylindrical specimens is described. Times of heating of specimens varied from 100 to 500 h; temperatures were within the limits 300–600° C. FeO, α-Fe_2O_3 and Fe_3O_4 films were tested; crystalline structures of the films are given. Tensile stresses needed for separating metallic specimens connected by films varied from 0·04 for FeO to > 4 for Fe_3O_4 in kg/cm².

Accelerated High Temperature Oxidation due to Vanadium Pentoxide. K. Sachs. (*Metallurgia*, 1958, **57**, Mar., 123–137; Apr., 167–173; May, 224–232). Corrosion of gas turbines running on residual oils containing vanadium is investigated. Laboratory tests are described on austenitic steels and other materials and also on other corrosive constituents. The mechanism of attack is examined and primary and secondary effects are evaluated. A number of tests with water injection and the addition of various constituents to a synthetic ash are then reported and the present position of the problem is stated. (41 refs.).

Open Air and Laboratory Investigations of the Resistance of Steel Products to Atmospheric Corrosion. L. Jassowicz. (*Prace Inst. Hutn.*, 1958, **10**, (4), 226–236). Trials over 3 years at 10 outside sites with various environments, one in a mine, and salt-box and humid SO_2 containing air trials in the laboratory are reported. Corten steel with 0·12%C, 0·47% Mn, 0·58%Si, 0·18%P, 0·021%S, 0·88%Cr, 0·12%Ni and 0·27%Cu proved the most resistant.

The Stress-Corrosion Cracking of Austenitic Stainless Steels in Aqueous Chloride Solution. T. P. Hoar and J. G. Hines. (*Nat. Metallurgical Lab., Council of Sci. and Ind. Res., Jamshedpur*, 1958, 414–424). Studies of 18–8 steels in hot $MgCl_2$ solutions are reported. After a period of induction a period of rapid crack propagation sets in. Increased stress or temperature shorten the induction period as does cold-work, but it may be lengthened by cathodic protection. A theory is advanced of selective anodic dissolution. **Effect of Low Alloy Additions on Corrosion of Steel in Different Atmospheres.** K. P. Mukherjee, A. K. Lahiri and T. Banerjee. (425–432). Studies on steels in industrial and marine atmospheres are reported showing the effects of Cu, Cr, Ni, Al, and other elements.

Stress-Corrosion Cracking of Austenitic Stainless Steels with Applied E.M.F. J. G. Hines and T. P. Hoar. (*J. Appl. Chem.*, 1958, **8**, Nov., 764–776). 18–8 Steels in the form of wire were tested in air-saturated 42%$MgCl_2$ solution at 150–154°. Mild cathodic polarization delayed cracking. After cracking has started, cathodic polarization retards it, and above a critical potential halts it. The method may be useful in practice.

Four Case Histories of Intergranular Corrosion of Austenitic Stainless Steel in Petroleum Refineries. J. R. Cunningham. (*Corrosion*, 1958, **14**, Dec. 121). Furnace tubes in a cracker, pipelines in a fixed bed catalytic reforming unit, linings in pressure vessels handling H_3PO_4 and a vessel containing organic acids are considered.

Stress Corrosion Cracking of Austenitic Steels at Elevated Temperatures and Pressures. V. P. Sidorov and A. V. Ryabchenkov. (*Metallov. Obra. Met.*, 1958, (6), 25–32). Three steels for boiler tubes and superheaters were tested in a device using the specimen as its own autoclave. Tests in NaOH solution and in pure condensate in presence of O_2 were carried out and a critical concentration of the solution was found for austenitic steels (varying with temperature and pressure). Addition of NaCl inhibits corrosion and time-to-failure is exponentially related to absolute temperature. Relief of internal stresses may prevent corrosion.

Failure of Type 316 Stainless Steel Autoclave Components. J. P. Hugo and L. G. Nel. (*Corrosion*, 1958, **14**, Dec., 553t–556t). Fittings of 316 steel in a Monel 1 gallon unstirred autoclave unexpectedly failed by stress corrosion. Chloride possibly corroded the metal and failure was by residual stress. Some degree of cold-working of the components before installation was indicated.

Chloride Stress Corrosion Cracking of Austenitic Stainless Steel: Effects of Temperature and pH. L. R. Scharfstein and W. F. Brindley. (*Corrosion*, 1958, **14**, Dec., 588t–592t). Overstressed U-bends of 304 and 347 stainless steels were exposed to chloride solutions at 165–200° F at pH 6·5–7·5 and 10·6–11·2. It is concluded that specific conditions must be present for cracking in these steels to occur.

The Mechanism of Cavitation. E. G. Richardson. (*Wear*, 1958, **2**, Nov., 97–106). The number and size of gaseous nuclei existing in liquids are measured. It is shown that they can be removed by long standing or by pressure. They are detected by the absorption of sound and the distinction between air nuclei and vapour nuclei is brought out.

Propeller Corrosion and Cavitation Erosion: Its Prevention by Cathodic Protection. M. G. Duff. (*Corros. Techn.*, 1958, **5**, Aug., 250–253). The author deals with the cathodic protection of steel ships with bronze propellers, steel ships with cast iron propellers and wooden vessels with bronze propellers. Position of anodes for each type of vessel is described and the theory of propeller breakdown discussed.—T. G.

Avoid Cavitation Damage. W. J. Rheingans. (*Mat. Design Eng.*, 1958, **48**, Sept., 102–106). Selection of stainless steels and other materials is discussed. The mechanism of cavitation is outlined and effects are shown.

Cavitation and Nuclei. R. T. Knapp. (*Trans. A.S.M.E.*, 1958, **80**, Aug., 1315–1324). Although liquids have high tensile strength cavitation appears as soon as the pressure drops below the vapour pressure, implying that the tensile strength is zero. This is explained by the presence of "weak spots." These are investigated and shown to be consistent with Harvey's theory of the nucleus. The engineering significance of the nuclei is discussed.

Procedure for Working Out the Heating Conditions for Steel on the Basis of Scale Thickness, Bark Thickness, and Grain Size. I. F. Galovnev. (*Metallov. Obra. Met.*, 1958, (1), 49–55). A range of steels was examined of which the scaling characteristics were known and after typical heat-treatment processes the scale was dissolved off and the underlying metal examined. Scaling is greatly affected by atmospheric humidity and 1% increase was found to increase scaling thickness by 1·5–5%. This is the explanation of the great variations in previously published results. A stainless steel gave a curve against humidity with a factor of 70 in scaling rate from 30%RH at 5° to 100% at 35°, and consequently a variation of thickness of about 1 : 8, with a parabolic law. Diagrams are given and the method of using them to predict scaling rates is explained.

Texture of Iron Scale. XII. Structure Variations in the Scale in Atmospheres Forming the Highest Oxides and in an Atmosphere Forming Wustite. V. I. Arkharov, M. I. Simonova and M. K. Yut. (*Fiz. Met.*, 1958, **6**, (3), 444–449). X-Ray investigations are reported on massive and tubular specimens into the changes of structure of iron scale forming in atmospheres in equilibrium for the formation of the higher oxides Fe_2O_3 and Fe_3O_4 and in an atmosphere in equilibrium for the formation of FeO. The reduction process of ferric to ferrous oxide develops on the outside and is accompanied by a diffusion of iron ions into the interior of the scale. The volume changes occurring during the reconstruction of the lattice of the higher oxides to the one corresponding to the lower oxide causes stresses in the decaying layer of the scale.—L. B.

Selective Oxidation of Al from an Al–Fe Alloy. R. E. Grace and A. U. Seybolt. (*J. Electrochem. Soc.*, 1958, **105**, Oct., 582–585). A 4–8%Al–Fe alloy was studied at 700–900° C in damp H_2 with dew-points from − 55° to 0°C. The product formed was γ-Al_2O_3. A parabolic law was followed and the activation energy was measured. (72 kcal/mole with 0° dew-point).

An Electrometric Method for Measuring the Thickness of the Air-formed Oxide Film on Pure Iron and Mild Steel. P. Hancock and J. E. O. Mayne. (*J. Chem. Soc.*, 1958, Nov., 4167–4172). The film is reduced cathodically and the thickness calculated from the quantity of electricity consumed. The method was confirmed gravimetrically. The thickness was 2·6–4 mμ depending on surface condition. **The Effect of Inhibitors of the Corrosion of Iron on the Air-formed Film.** (4172–4175). 0·1 N Solutions of Na benzoate, acetate, borate, carbonate and hydroxide were used. The film was found to be increased in thickness by 20–120% by material of similar composition depending upon the inhibitor and the condition of the surface.

Corrosion Research and its Industrial Background. W. H. J. Vernon. (*Chem. Ind.*, 1958, Oct. 25, 1381–1389). A lecture. A considerable historical review is presented with a discussion of the economic importance of corrosion processes. (27 refs.).

Corrosion Trials at Present in Use for the Study of Electro-Cast Refractory Products. P. Bortaud. (*Silicates Ind.*, 1958, **22**, Sept., 443–446). The equipment and technique used for accelerated corrosion tests on electro-cast refractories are described, together with a modification for testing iron and steelmaking applications.

Influence of Nickel on the Electrochemical and Corrosion Behaviour of Steel. G. M. Florianovich, Y. M. Kolotyrkin and N. K. Smirnova. (*Doklady A.N.*, 1958, **120**, (4), 845–848). Using a potentiostatic method of measurement of polarization the influence of nickel on the behaviour of steel in the region of passivation was studied. Measurement was carried out in a 0·1N solution of H_2SO_4 in a nitrogen atmosphere at 20° C. Details of this work are discussed and it is concluded that the degree of passivity of the surface of steel is a function of the potential and does not depend on the method of its maintenance.—S. I. T.

On Diagnosing the Causes of Corrosion from the Products of Corrosion. G. Schikorr. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 552–556). Corrosion is considered as a disease of metals. Symptoms are listed under the heads: composition of products of corrosion and distribution of corrosion products. It seems that corrosion products may be distributed in four different ways, each of diagnostic significance: (1) all the products are on the metal affected; (2) a substantial part of the products is not on the metal affected (e.g. paper wrappings of steel may have most stains); (3) the metal is practically free from corrosion products (e.g. steel pipes in brickwork); (4) the products are unevenly distributed over the metal. The significance of these symptoms is discussed.—L. J. L.

Measurement of Inter-Crystalline Corrosion by the "Resonance Frequency" Method. K. Smrček. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 550–552). The apparatus described makes use of the fact that the resonance frequency of a metal specimen is affected by intercrystalline corrosion.—L. J. L.

Amine Nitrites in the Protection of Iron Against Oxidation. M. P. Monserrat and F. E. Prosper. (*Rev. Ciencia Apl.*, 1958, **12**, (4), 293–298). The preparation and properties of the nitrites of dicyclohexylamine and triethanolamine are described, and their uses as corrosion inhibitors discussed.

Mechanism of Inhibiting Effect of Hydrofluoric Acid in Fuming Nitric Acid on Liquid-Phase Corrosion of Aluminium and Steel Alloys. D. M. Mason and J. B. Rittenhouse. (*Corrosion*, 1958, **14**, July, 345t–347t). A tenacious insoluble fluoride is formed in both cases.

The Activation and Inhibition of Corrosion of Metals by Metallic Cations. W. R. Buck and H. Leidheiser jun. (*Corrosion*, 1958, **14**, July, 308t–312t). The effects of other metals on corrosion by fruit juices are examined. Other metal pairs are also tested.

Basic Research into the Corrosion of Steel in Salt Water. (*Corros. Techn.*, 1958, **5**, Sept., 295–296). An outline of work recently completed by the U.S. National Bureau of Standards to determine the electric potential required to protect steel cathodically from salt water. It was found that, under controlled laboratory conditions, the best protection was obtained at − 0·72 V vs. the standard calomel electrode.

What You Can Do to Reduce Stress Corrosion. F. J. Poss. (*Chem. Eng.*, 1958, **65**, July 28, 140, 142, 144). The article refers to austenitic stainless steels. General principles are indicated.

Corrosion and Passivity. V. F. Franck. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 504–513). A survey is presented of the present state of knowledge on the passivity of metals with particular reference to protection against corrosion. The survey is based on the papers presented at the colloquium on the passivity of metals, held at Darmstadt in September, 1957, under the auspices of the American Electrochemical Society, the Faraday Society, and the German Bunsengesellschaft.—L. J. L.

A New Angle on Corrosion-Protection. K. Thalhofer. (*Werks. Korr.*, 1958, **9**, Aug.-Sept., 515–517). Nine different kinds of corrosion are listed; existing methods of providing protection against corrosion are critically examined and a new approach to the problem of protecting constructional materials is advocated. It may have to start with the winning of the raw materials, and should then be extended to preparation and manufacture, to transport to the site, and to erection.

Corrosion and Water Treatment. G. G. Sindery. (*Corros. Techn.*, 1958, **5**, July, 209–212, 224). The author first discusses, and indicates the applications of the normal water pretreatment methods consisting of precipitation and ion-exchange. The author then considers the best conditioning treatments and the various suitable corrosion inhibitors for use in high and low pressure boilers, condensate systems, water-cooling and hot-water systems, and cold-water lines.

Water Treatment for Domestic Hot-Water Supplies. C. C. Hanson. (*Corros. Techn.*, 1958, **5**, July, 220–224). The author discusses the nature of scale formation and the corrosion of iron pipes and shows how the use of sodium hexametaphosphate minimises deposition and corrosion.—L. E. W.

The Role of Corrosion Inhibitors in Water Treatment. E. L. Streatfield. (*Mét.-Corr.-Ind.*, 1958, Oct., 420–430). Softening processes are outlined, especially by exchange, and corrosion by deionised water and the effects of anodic, cathodic and such organic inhibitors as sodium benzoate and amines are described.

Theory and Practice of Boiler Water Treatment as a Means of Eliminating Corrosion and Scale. E. Quaade-Holm. (*Corros. Techn.*, 1958, **5**, July, 213–214, 237). The author lists the principal causes of corrosion in steam boilers as (1) the presence of atmospheric oxygen in the water, (2) differential aeration and (3) the action of free acids. The prevention of scale by chemical treatment depends upon a perfect stoichiometric balance with regard to the quality and quantity of the various reagents with which the feed-water is treated. In order to maintain such balance, laboratory controls are needed. There is a simpler and cheaper physical treatment based upon organic components (having mol. wts. about 30 times greater than those of dissolved salts) which absorb both the positive and negative ions formed in the water. The cations and anions neutralize each other and are then precipitated, which eliminates the possibility of scale formation. The author concludes by discussing the removal of old scale.—L. E. W.

Magnetic Treatment of Liquids for Scale and Corrosion Prevention. T. Vermeiren. (*Corros. Techn.*, 1958, **5**, July, 215–219). The treatment modifies the conditions of separation of crystalline matter by favouring the formation of matter in suspension. The method consists of circulating the liquid to be treated through two annular gaps between the extremities of a powerful magnet and the outer body, i.e. where the air-gap is smallest and the field strongest. The author analyses the causes of failure and discusses the chemical effects (e.g. increased pH), of the method.—L. E. W.

Cathodic Protection of Radiant Heating Pipes. L. C. Wasson. (*Corrosion*, 1958, **14**, July, 333t–336t). In the case described a ground bed could not be used and anodes were installed through a concrete floor. About 500 anode clusters were placed in an area 300 × 1200 ft, and the techniques employed are described and cost data are given.

Use of Magnesium Anodes to Protect Steel Mains. H. C. Boone. (*Corrosion*, 1958, **14**, July, 307t). Protection of a 30-in. cast-iron pipe replaced in part by a 20-in. steel main, is outlined.

New Corrosion Research Laboratory. (*Corros. Techn.*, 1958, **5**, Nov., 361–362). A description of the new corrosion research laboratory opened by Carpenter Steel Co., of Reading, Pa. Important new means of determining accurately the corrosion rate on metals have been introduced.

New Ways to Look at Corrosion Processes. E. A. Gulbransen. (*Chem. Eng. Prog.*, 1958, **54**, Nov., 60–61). The approach by electron diffraction study of oxides formed is briefly described. Stress-corroded stainless steel is examined.

A Coating Evaluation Testing Program. E. R. Allen, jun. (*Corrosion*, 1958, **14**, Dec., 546t–552t). Pipeline coatings were tested by field tests begun 10 years ago and are now correlated with laboratory trials. Procedures and apparatus were developed, coatings in use were tested and then new materials were examined and compared with those in use.

Sulphuric Acid Corrosion of Stainless Steels. W. L. Mathay. (*Ind. Eng. Chem.*, 1958, **50**, Sept. part I, 85A–86A). Tests in plant equipment should be made. Various factors in the resistance of AISI 316, 321 and 347 types are discussed.

Effect of Temperature Variation on Composition, Fouling Tendency, and Corrosiveness of Combustion Gas From a Pulverized-Fuel-Fired Steam Generator. J. D. Piper and H. Van Vliet. (*Trans. A.S.M.E.*, 1958, **80**, Aug., 1251–1263). Metal condensers maintained at temperatures between 87–242° F were placed in stack gas and deposits as well as degrees of corrosion were observed. Corrosion rates increase as the dew-point is approached.

A Polarographic Study of the Corrosion of Iron and Some Ferrous Alloys by Sulphur Dioxide. F. J. Bowen and A. H. Gropp. (*Corrosion*, 1958, **14**, Dec., 578t–582t). Pure iron and 7 alloy steels were tested for periods of 12 h. The consumption of SO_2 was determined. A mechanism of corrosion yielding sulphide is discussed.

Effect of Increasing the Area of Iron to Zinc on the Rate of Reversal of Potential in a Zinc-Iron System. H. L. Shuldener and L. Lehrman. (*Corrosion*, 1958, **14**, Dec., 545t). Iron tends to become anodic to zinc in water systems. Increase of the ratio of iron area to Zn area hastens potential reversal.

The Sulphur Dioxide Test. J. Edwards. (*Met. Fin.*, 1958, **56**, Oct., 49–55). An accelerated corrosion test for steels coated with Cr, Ni and other metals or combinations of metals using an atmosphere with a controlled SO_2 content is described, though the content of SO_2 was not found to be critical. Relative humidity is important, attack being less rapid below 95%.

Measuring Corrosion by the Electrical Conductance Test Method. E. Ford. (*Corros. Techn.*, 1958, **5**, Oct., 330–331). By noting changes in electrical conductance, the reduction in cross-section of a corroding specimen may be evaluated. Advantages of this method are discussed and its application to such problems as the relation between steel corrosion and humidity is suggested.—A. G.

Investigation of Erosion and Corrosion of Turbine Materials in Wet Oxygenated Steam.

H. A. Cataldi, C. F. Cheng and V. S. Musick. (*Trans. A.S.T.M.*, 1958, **80**, Oct., 1465–1478). Laboratory tests on various materials for use in the steam turbines of the Dresden nuclear power station fed with $\frac{1}{2}$ % moisture, rather high O_2 and 869 psi/525° F are reported. Except for a few critical locations, conventional materials are suitable.

An Index of Cavitation Erosion by Means of Radioisotopes. S. L. Kerr and K. Rosenberg. (*Trans. A.S.M.E.*, 1958, **80**, Aug., 1308–1314). Tests on a hydraulic turbine painted with a copal-base paint containing ^{76}As are described.

Study of Corrosion and Cavitation-Erosion Damage. J. Z. Lichtman, D. H. Kallas, C. K. Chatten and E. P. Cockran, jun. (*Trans. A.S.M.E.*, 1958, **80**, Aug., 1325–1341). Damage to ships was investigated and work in water tunnels using a magnetostrictive transducer, a piezoelectric transducer and a rotating disc apparatus was carried out. Cavity collapse is shown to be confined to propellers. Damage to struts, rudders and other appendages is primarily due to corrosion aggravated by mechanical scouring.

The Mechanism of Passivating-Type Inhibitors. M. Stern. (*J. Electrochem. Soc.*, 1958, **105**, Nov., 638–647). Local current action anodically polarizes the metal giving a noble mixed potential. This occurs whether oxides or adsorbed films are formed. Factors in protection are discussed and the special cases of O_2, chromate, molybdate and pertechnetate on stainless steel and Ti are considered.

Evaluation of Refinery Corrosion Inhibitors. A. J. Freedman and A. Drawnieks. (*Corrosion*, 1958, **14**, Dec., 567t–570t). Laboratory tests have been developed to simulate certain specific environmental factors occurring in field conditions. Work on inhibitors for light naphtha condensers, distillation towers, special high-temperature locations, high velocity service, narrow crevices and areas exposed to impingement attack is reported. Electrical resistance probe trials confirm the findings.

Corrosion Control Courses: What Colleges are Teaching. R. V. James. (*Corrosion*, 1958, **14**, Dec., 122, 124). Compiled from a questionnaire to 103 engineering colleges in U.S.A.

How Environment Directs Corrosion Control. R. V. Jelinek. (*Chem. Eng.*, 1958, **65**, Sept. 22, 163–168). Effects of pH, oxidizing agents, temperature and stream velocity are outlined and the action of inhibitors is considered with a table of these agents and their applications. Cathodic protection is mentioned. (20 refs.).

The Prevention of Air Heater Corrosion. L. E. Hardy. (*Steam Eng.*, 1958, **28**, Oct., 20–21). Methods of reducing corrosion when firing boilers with oils of high sulphur content, in the low-temperature zone of air heaters.

Protecting Steel Against Wind and Wave. (*Esso Oilways*, 1958, **9**, (5), 21–24). The work of the Danish Central Paint and Varnish Laboratory on protective paint systems is outlined.

Protection of Iron by Rust Stabilisation. (*Ind. Fin.*, 1958, **10**, Sept., 47). A primer converting iron hydroxides to Fe_3O_4 is described and weathering and deep drawing tests are quoted.

Cathodic Protection. L. B. Hogben. (*Elect. Rev.*, 1958, **163**, Nov. 28, 992). A summary of a paper read to Incorporated Plant Engineers. Underground corrosion conditions and protection systems are briefly outlined.

Criteria for Adequate Cathodic Protection of Coated, Buried or Submerged Steel Pipe Lines and Similar Steel Structures. NACE Tech. Unit. Committee T-2C. (*Corrosion*, 1958, **14**, Dec., 561t). The use of a protective current sufficient to maintain the external surface at every point negative by at least 0·85V to $Cu/CuSO_4$ half-cell is recommended subject to reservations.

New Studies in the Cathodic Protection of Gas Mains. (*Gas Times*, 1958, **92**, Sept., 38–39, 41). An account of the course on "Underground Corrosion and Cathodic Protection" given at the Summer School of Battersea College of Technology in July.

Zinc Alloy Anodes. J. T. Grennell and W. C. G. Wheeler. (*J. Appl. Chem.*, 1958, **8**, Sept., 571–576). Additions of Al ($\frac{1}{2}$–1%) and of Si ($\frac{1}{2}$–$\frac{1}{4}$%) are shown to overcome the detrimental effect of iron and to give improved performance. The iron is not dissolved in cold dilute H_2SO_4 when bound by Si and the amount dissolved is a measure of the "harmful" iron present.

Lead Alloy Anode for Cathodic Protection. J. H. Morgan. (*Corros. Techn.*, 1958, **5**, 347–352). The mode of operation of a corrosion product impressed-current anode and the formation of a conducting oxide film on a Pb alloy are outlined. Some designs and performances are reviewed.

ANALYSIS

Photometric Determination of Iron (III) Employing 1 : 2-*cyclo*-Hexanediaminetetraacetic Acid. F. Bermejo Martínez and R. Rey Mendoza. (*Chem-Analyst*, 1958, **47**, Dec., 94–95).

pH Adjustment in Colorimetric Iron Determinations. N. F. Davis, C. E. Osborne, jun. and H. A. Nash. (*Anal. Chem.*, 1958, **30**, Dec., 2035). Using 2:2'-dipyridyl and *o*-phenanthroline the reagent should be added before pH adjustment.

Rapid Analysis of Carbon in Molten Steel by Means of Induction Heating and the Dry CO_2 Absorbing Method. D. Kamimori. (*Tetsu to Hagane*, 1958, **44**, Mar., 247–249).—K. E. J.

The Direct Determination of Combined Carbon in Grey Pig and Cast Iron. H. Krapp and K.-H. Tytko. (*Giesserei*, 1958, **45**, Dec. 18, 778). Some comments on the discussion of the article with the above title, 1958, (21), 639–642; (22), 671).

Recommendation for the Elimination of the "Dead Space" in the Apparatus for the Determination of Sulphur in Steel and Iron by the Schulte-Franke Method. W. Schluttig. (*Giessereitechn.*, 1958, **4**, Nov., 294). Details of modifications in the apparatus are given.

Determination of Si, Mn, P, Cr, and Mo Contents in Pig Iron or Cast Iron from a Single Solution. L. Egri. (*Koh. Lapok-Öntöde*, 1958, **9**, August, 194–196). A rapid, exact and cheap method for the analysis of a single solution of pig iron or cast iron is described employing gravimetric analysis for Si, titrimetric analysis for Mn, and photometric analysis for P, Cr and Mo.—P. K.

Determination of a Small Amount of Boron in Steel by the Curcumin Method. A. Nimura. (*Tetsu to Hagane*, 1958, **44**, Mar., 253–255).—K. E. J.

Qualitative Analysis and Colorimetric Determination of Cobalt by Means of Carboxy-2-isonitrosoacetanilide. F. Buscarons and L. Ma. Munné. (*Anal. Chim. Acta*, 1958, **19**, Nov., (5), 432–434). [In French]. By means of this reagent, *N*-isonitrosoacetylanthranilic acid, which is specific for Co, cobaltic ion can be detected down to 1 in 10^6. Conditions are given for the colorimetric determination of Co from the wine-red colour produced, masking Ni with cyanide.

Determination of Aluminium in Constructional Steel. E. Hekneby. (*Tidsskr. Kjemi Bergv. Met.*, 1958, **18**, (9), 148–150). A two-stage electrolytic separation using a Hg cathode and a gravimetric determination with cupferron is described.

Photometric Determination of Aluminium in Steels after Separation by Means of Ion-Exchange Resins. M. L. Foglino and G. P. Spagliardi. (*Met. Ital.*, 1958, **8**, Aug., 372–374). The method of determining Al in steels described by the authors makes use of Amberlite IRA-400-OH resin made by Rohm and Haas. Cr is volatilized and thioglycollic acid is added to help agglomerate all traces of Fe not retained by the resin. The method is claimed to be both rapid and accurate.—M. D. J. B.

A Rapid Method for the Determination of Lead in Free-Cutting Steel. S. Meyer and O. G. Koch. (*Arch. Eisenhütt.*, 1958, **29**, Nov., 677–681). A description is given of three methods of determining the lead content of steel. The lead is separated in the form of sulphate and is determined gravimetrically as molybdate or chromate, and potentiometrically as chromate.—L. J. L.

Analysis of Microquantities of Antimony, Tin and Arsenic in Ferro-Tungsten. H. Goto, Y. Kakita and M. Sase. (*Nippon Kinzoku*, 1957, **21**, June, 385–387). A spectrophotometric method depends on separation of the elements as sulphides, followed by determinations using molybdenum blue, methyl violet-amyl acetate, and a cacothelin method, respectively.—K. E. J.

Spectrophotometric Methods for the Analysis of Steels. A. Devoti and A. Sommariva. (*Met. Ital.*, 1958, **8**, Aug., 355–366). The authors describe spectrophotometric methods of determining Al, Cr, P and Si in steels. The Beckman D.U. apparatus used is also described and considerable technical and statistical data are given as well as a comprehensive bibliography.—M. D. J. B.

Ultra-High Dispersion Spectrographs and Their Use in the Investigation of Iron and Steel. G. Graue, S. Eckhard and R. Marotz. (*Arch. Eisenh.*, 1958, **29**, Oct., 619–625, discussion 626). A description of a high-dispersion grating spectrograph is given, and its advantages in the resolution of complex spectra discussed, with special reference to Fe spectra.

On Routine Analysis by Means of the Quantorecorder. K. Ando. (*Tetsu to Hagane*, 1958, **44**, Mar., 249–251). For various steels, comparisons are made between chemical and Quantorecorder analyses for C, Si, Mn, Cr, Ni, Mo, Cu and P. The effect of C on the results for Mn and Cr is discussed. A report of an analysis can be produced within 5 min of taking the sample.—K. E. J.

Spectrographic Analysis of Structural Steels by the Plane Point Technique with a Copper Counterelectrode. P. Murtas. (*Met. Ital.*, 1958, **8**, Aug., 375–378). The need to carry out tests on small finished parts, such as bolts, screws and dies, too small to offer adequate area for striking a spark, has resulted in a new process known as the plane/point method with copper counterelectrode. This technique was applied to Fe–Cr, Fe–Mn and Ni–Cr alloys by N. N. Sorokina. The technique has the advantages that it can be applied to the analysis of very small pieces and that it uses electrolytic copper electrodes in place of costly graphite electrodes. An intermittent high potential medium inductance spark is used.—M. D. J. B.

Spectrophotometric Determination of Combined Carbon in Cast Iron. M. L. Foglino and G. P. Spagliardi. (*Met. Ital.*, 1958, **8**, Aug., 381–383). The authors discuss the relative merits of the experimental methods of Newberg and Jamieson for the determination of combined carbon in ingot mould irons and in nodular irons. Technical and statistical data are given.—M. D. J. B.

Simultaneous Spectrographic Analysis of Carbon and Other Alloying Elements in Steels and Brasses. A. S. Dem'yanchuk. (*Avtom. Svarka*, 1958, (8), 41–45). A method is described for the simultaneous determination of the contents of C and other additions to ordinary carbon and low alloy steels and also to brasses by means of high-frequency discharge emission as a source of spectral excitation. Simultaneous determination of different additions saves time. The method is very consistent and accurate.—M. T.

Spectrographic Determination of Boron and Silicon in Low-Alloy Steel by Fluoride Evolution. J. E. Paterson and W. F. Grimes (*Anal. Chem.*, 1958, **30**, Dec., 1909–1902). CuF_2 is used as a source of F and the metal is converted to the mixed oxides by HNO_3 treatment (other acids give variable B recovery) and examined in a deep-cratered supporting electrode.

Determination of Small Amounts of Lead in Electrolytic Manganese and Ferromanganese. H. Goto and K. Hirokawa. (*Nippon Kinzoku*, 1957, **21**, Jan., 65–68). The method is spectrophotometric, using Dithizone. With ferromanganese, lead is extracted by Dithizone from the sample solution and determined separately. (14 references).—K. E. J.

X-Ray Spectrophotographic Determination of Tantalum, Niobium, Iron and Titanium Oxide Mixtures. B. J. Mitchell. (*Anal.*

Chem., 1958, **30**, Dec., 1894–1900). Corrections for absorption and inter-element effects are given.

Spectrophotometric Determination of Tin with Quinalizarin. S. Wakamatsu. (*Nippon Kinzoku*, 1957, **21**, July, 450–453). Tin in iron and steel can be determined spectrophotometrically using the reaction of quinalizarin with the stannic ion in ammonium acetate buffer at pH 3·0–3·3. The method requires 30 min after dissolution of the sample.—K. E. J.

A New Spectrophotometric Determination of Small Amounts of Titanium in Steel and Ferro-Alloys. I.—Determination of Titanium in Steel and Cast Iron. H. Goto, Y. Kakita and K. Hirokawa. (*Sci. Rep. Res. Inst. Tohoku Univ.*, 1958, **10A**, Oct., 356–360). [In English]. 0·002–0·2%Ti in steel and cast iron is determined absorptiometrically by the alizarinsulphonate method. The accuracy is high. Iron does not interfere, but Cr does in excess of 20% in steels. **II.—Determination of Titanium in Metallic Ferro-Manganese, Metallic Silicon, Ferro-Silicon and Silico-Manganese. III.—Determination of Titanium in Metallic Chromium, Ferro-Chromium, Ferro-Tungsten, Ferro-Molybdenum and Ferro-Vanadium.** H. Goto and K. Hirokawa. (361–366; 367–374). II.—Ti is determined absorptiometrically by the alizarinsulphonate method. Mn does not interfere, but Si has to be removed with HF. III.—Ti is determined absorptiometrically by the alizarinsulphonate method. Cr, W, Mo, and V interfere and must be separated. The accuracy of the determination is high.

Determination of Tungsten or Molybdenum by X-Ray Emission Spectrography. J. E. Fagel, jun., H. A. Liebhafsky and P. D. Zemany. (*Anal. Chem.*, 1958, **30**, Dec., 1918–1920). Alkaline sodium tungstate solution is used with NaBr as internal standard.

The Spectrographic Method for the Determination of Magnesium in Spheroidal Graphite Iron. E. Melgara. (*Met. Ital.*, 1958, **8**, Aug., 378–380). The method of determining Mg in spheroidal graphite iron by spectrographic analysis is described and a brief review is made of the nature of the samples used, the apparatus, photographic equipment and the analytical procedure. Results of the investigation are discussed and technical data are given in table form.—M. D. J. B.

Spectrophotometric Determination of Vanadium in Steels. G. Picasso. (*Met. Ital.*, 1958, **8**, Aug., 367–372). The author describes the spectrophotometric method of determining V in steels and discusses apparatus, solutions and reagents used. The effects of ammonia and acetone are examined on colour formation in the V (III) system. The effects of V valency, of acidity during reduction by $SnCl_2$, of sulphuric acid concentration during colour development and the interference due to the presence of various ions are also discussed. Recommendations are put forward for normal routine analysis and also for the precision analysis of steels.—M. D. J. B.

Ferrous Metallurgy: Review of Applied Analysis. H. F. Beeghly. (*Anal. Chem.*. 1959, **31**, Apr. Part 2, 706–712). A general review with 182 refs.

Gases in Iron and Steel. C. E. A. Shanahan. (*Iron Coal Trades Rev.*, 1959, **178**, Apr. 10, 845–851). The behaviour of Ni, H_2 and O_2 in iron and steel and their determination, especially of O_2 by a new procedure, are discussed.

Rapid Hydrogen Determination for Steel Foundry Control. C.C. Carson and B.J. Alperin. (*Mod. Castings*, 1959, **34**, Feb., 50–74). The method for the determination of H_2 in steel is rapid, simple and economical. It uses the hot extraction technique and employs a Pirani gauge for the analysis of the extracted gases. The method is described in detail.—T. G.

Fast Oxygen Analysis Speeds Output. (*Steel*, 1959, **144**, Mar. 2, 124–127). The components and operation of the conductometric oxygen analyser are described. The instrument can analyse a 1 gm sample in 5–8 min with an accuracy of ± 0·0002%.—A. G.

Combustion Method for Determination of Sulphur in Ferrous Alloys: Modifications for Minimizing Errors. J. W. Fulton and R. E. Fryxell. (*Anal. Chem.*, 1959, **31**, Mar., 401–405). A tracer study shows surface adsorption of SO_2 to be the primary source of error. Modification of the apparatus and titration with alkali improve precision.

Alkalimetric Determination of Phosphorus as Quinoline Phosphomolybdate. G. Alfelt. (*Jernkont. Ann.*, 1959, **143**, (3), 167–174). Methods for steel are given. The pH has a considerable effect on the precipitation and the amount of reagent required.

The Determination of Cobalt, Iron plus Vanadium or Manganese in Soft Magnetic Alloys. D. H. Wilkins and L. E. Hibbs jun. (*Anal. Chim. Acta.*, 1959, **20**, May, 427–429). An ion-exchange separation and an EDTA back titration are used.

Rapid Determination of Niobium in Stainless Steel. R. St. J. Emery. (*Metallurgia*, 1959, **59**, Feb., 101–103). Nb is determined spectrophotometrically in 18–8 stainless steels in 90 min using thiocyanate. Optical density is measured at 383 mμ, and a Spekker may be used.

Study of Silicon Nitride in steel. I: Determination of Silicon Nitride Nitrogen. H. Sawamura, T. Mori, K. Fujita and H. Miura. (*Suiyokwai-Shi*, 1958, **13**, Dec., 681–684). The ester-halogen residue was fused with NaOH and the NH_3 absorbed in N/10 HCl. The X-ray powder pattern suggested Si_3N_4 in steels with 0·39% and 3·09%Si.

Determination of Vanadium in Ferro-Vanadium. B. E. Jaboulay. (*Chim. Anal.*, 1959, **41**, May, 185). A modification of a method previously suggested. Samples of 0·35–0·25 g depending on V content are dissolved in $H_2SO_4 : HNO_3 : H_2O = 20 : 2 : 25$ (in ml). and the solutions oxidized with $KMnO_4$. H_3PO_4 is then added and titration is carried out with ferrous sulphate using Ba diphenylamine-sulphonate as indicator.

Separation of Rare Earths from Beryllium, Magnesium, Zirconium, Titanium, Uranium and Stainless Steel. M. W. Lerner and L. J. Pinto. (*Anal. Chem.*, 1959, **31**, Apr., 549–551). A recovery of at least 96% of the rare earth is obtained.

Colorimetric Determination of Aluminium in Heat-Resistant Alloys. H. Zibulsky, M. F. Slowinski and J. A. White. (*Anal. Chem.*, 1959, **31**, Feb., 280–281). Cupferron–$CHCl_3$ separation is followed by determination with 8-hydroxyquinoline.

Photometric Determination of Aluminium in Steel. U. T. Hill. (*Anal. Chem.*, 1959, **31**, Mar., 429–431). Nitrated Eriochrome Cyanine R is used, ascorbic acid is a good masking agent in high Fe concentrations.

Porous Cup Technique in the Determination of Magnesium in Cast Iron: A Statistical Study. A. C. Ottolini. (*Anal. Chem.*, 1959, **31**, Mar., 447–449). The range from 0·01–0·16% is covered.

Sulphate Determination in Chromium Plating Solutions. S. Konishi. (*Met. Fin.*, 1959, **57**, Feb., 59–60). A back titration method for $BaSO_4$ with E.D.T.A. is described.

Notes on Analysis of Sulphate in Chromium Plating Baths.† J. Ducey and G. McKinley. (*Plating*, 1959, **46**, Apr., 384).

The Microdetermination of Dissolved Oxygen by Hypodermic Syringe. P. Holland. (*Chem. Ind.*, 1959, Feb. 14, 218–219). A cheap and simple method using *o*-tolidine and as good as the sensitivity of the indicator (0·003 ppm) is described. A more sensitive end-point method would probably be justified.

Galvanic Analysis: Continuous Determination of Traces of Oxygen and Hydrogen. P. Hersch. (*Chim. Anal.*, 1959, **41**, May, 189–197). O_2 is determined directly and H_2 indirectly by combustion in excess O_2 by the effect on the Ag/KOH/Pb cell. An application to the H_2 evolved in corrosion is described.

The Spectrometric Determination of Carbon in Metals under Controlled Atmospheres. R. Breckpot, J. Morris and K. de Clippeleir. (*Rev. Univ. Mines*, 1959, **15**, May, 266–270). [In English]. In a H_2 atmosphere the accuracy of the determination of low concentration in iron is improved and the determination in Co is made possible.

Emission Spectrometric Determination of the Gaseous Elements in Metals. V. A. Fassel, W. A. Gordon, R. J. Jasinski and F. M. Evens. (*Rev. Univ. Mines*, 1959, **15**, May, 278–281). [In English]. A method of extracting volatiles in the d.c. arc and then exciting emission spectra with the same discharge is described. O_2 and N_2 in steels and O_2 in Zr, Ti, Th etc. can be determined.

Experience with Vacuum Spectrometric Analysis of Carbon, Sulphur and Phosphorus in Steel. S. Eckhard, G. Graue and R. Marotz. (*Rev. Univ. Mines*, 1959, **15**, May, 270–275). [In German]. Use of the Hilger vacuum fluorite spectrograph is described.

Recent Results Obtained with the Quantovac in Steel Analysis. W. H. Barry and J. M. Carrol. (*Rev. Univ. Mines.*, 1959, **15**, May, 281–289). Technique is described and precision discussed for determination of C, S, P, Mn, Si, Ni, Cr, Cu, Al and Mo.

On the Emission of CIII 2296·86 from the Different Spark Zones and a New Method of Determination of Carbon in Steel. T. Török and G. Szikora. (*Rev. Univ. Mines*, 1959, **15**, May, 275–278). [In German].

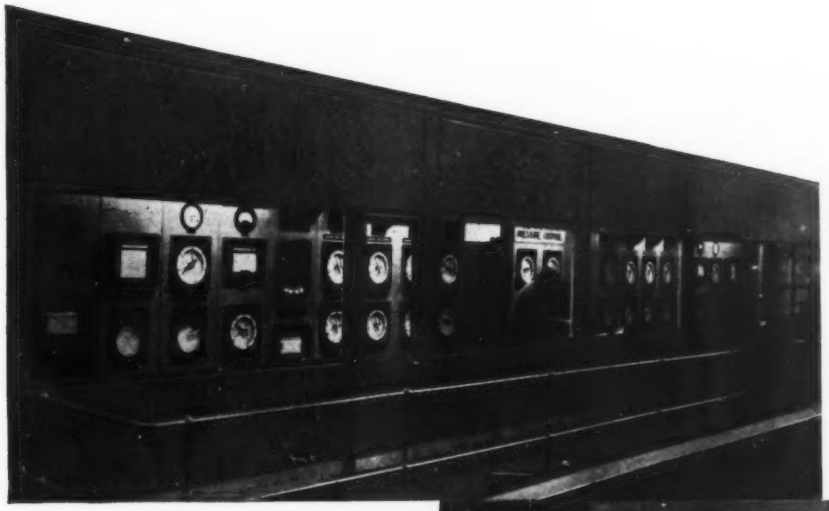
Spectrophotometric Determination of Cobalt with 1-(2-Pyridineazo)-2-naphthol : Separation from Interfering Ions. G. Goldstein, D. L. Manning and O. Menis. (*Anal. Chem.*, 1959, **31**, Feb., 192–195). An anion exchange separation, or citrate masking, removes all interfering elements but Cu. If present Cu can be removed electrolytically.

Spectrographic Analysis of Tungsten Metal Powder. R. Dyck and T. J. Veleker. (*Anal. Chem.*, 1959, **31**, Mar., 390–392).

Spectrographic Determination of Silicon, Aluminium and Magnesium in Limestones. Z. Gregorowicz. (*Chemia Anal.*, 1958, **3**, (5/6), 783–788). A method using Ca as internal and Cu and Fe as external standards is described.

Spectrographic Method for Dolomite Analysis. B. Tucholka-Szmeja. (*Chemia Anal.*, 1958, **3**, (5/6), 761–773). Simultaneous determination of Si, Al, Fe, Mg, Ca, Mg and Ti with a quartz spectrograph is described.

A New Method for Dissolving Aluminosilicates. S. Rossol. (*Chimia Anal.*, 1958, **3**, (5/6), 865–869). A mixture of HCl, HNO_3, H_2SO_4 and $HClO_4$ is used, mainly the two latter. (8 ml and 15 ml respectively for 0·5 g samples). The time taken, with two separations of SiO_2 is ~ 8 h. The method is simpler than the Na–K carbonates fusion and takes only one-third of the time. Other minerals and ores can be dissolved by the same process.



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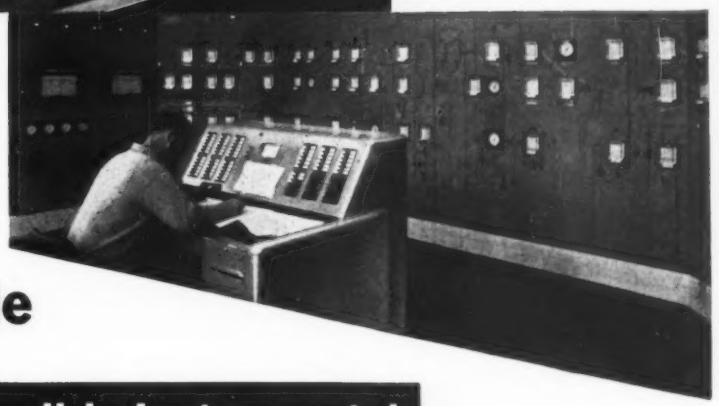

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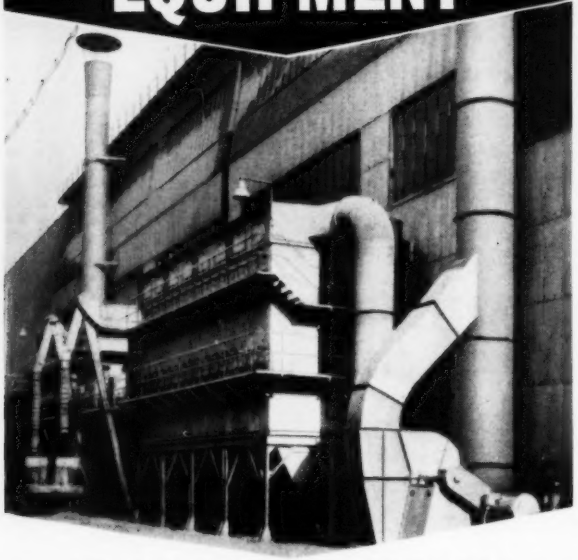

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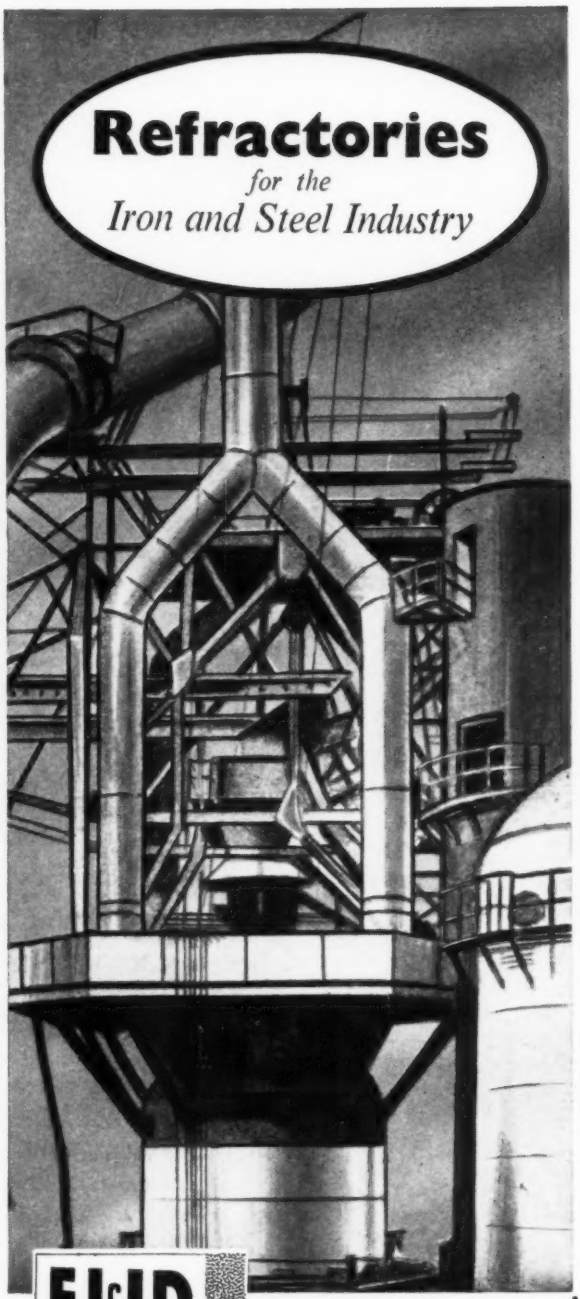

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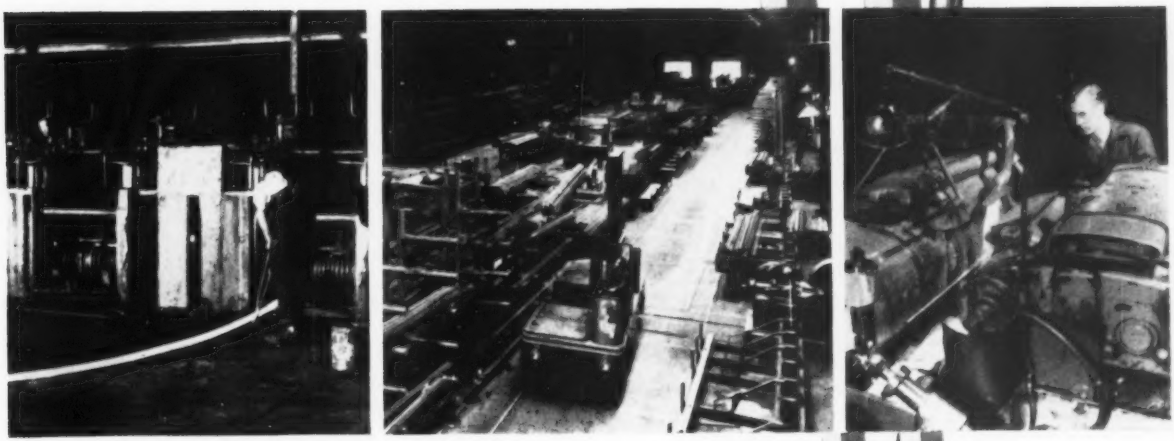

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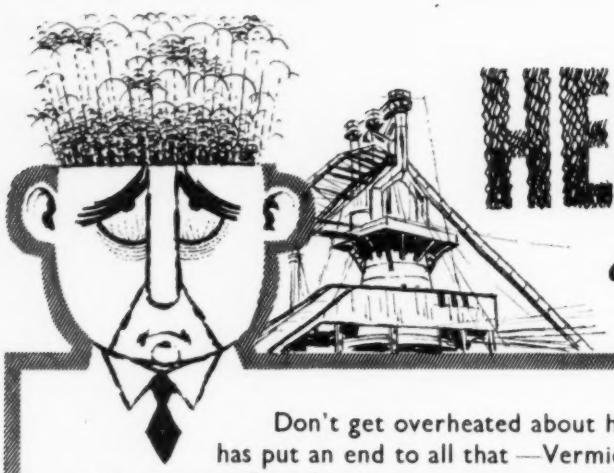

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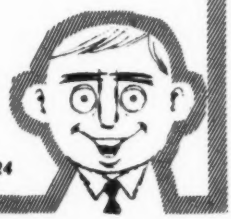

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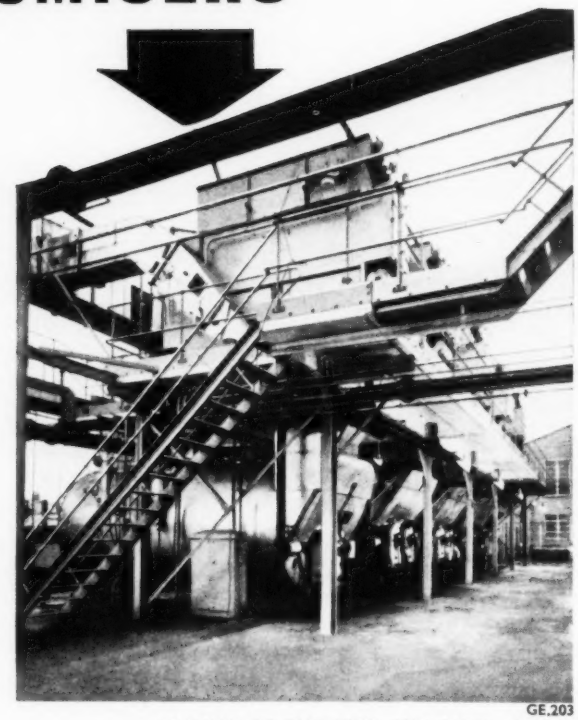

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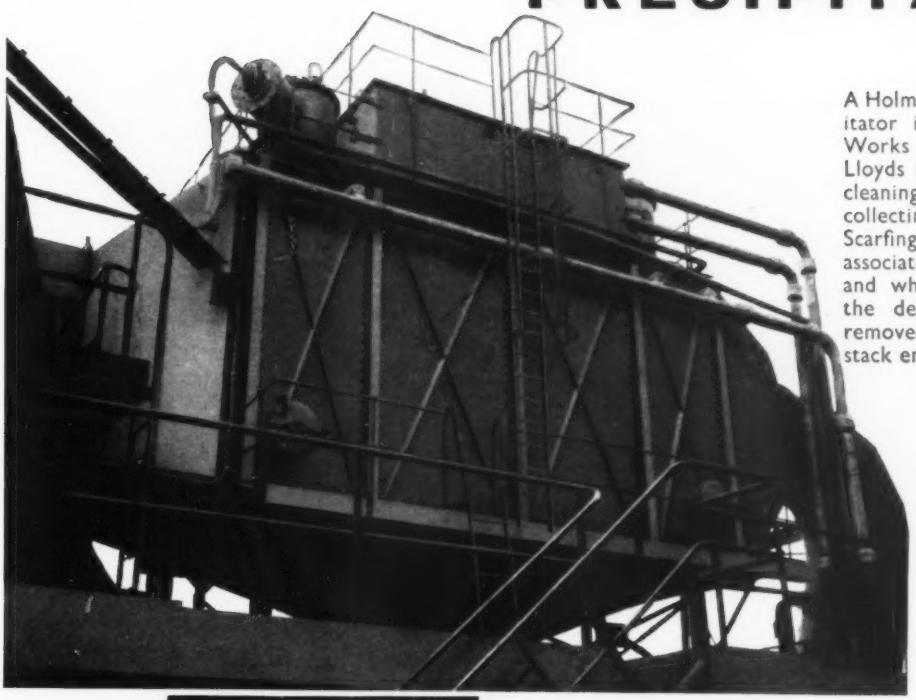

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The first edition of this survey appeared in 1951 as No. 44 in The Iron and Steel Institute's Special Report Series, and was quickly sold out. In view of the many requests for copies of what is now a standard work of reference, it was decided to issue a second edition, with some important additional material.

Through the courtesy of Richard Thomas and Baldwins Ltd., The Steel Company of Wales Ltd., and John Summers and Sons Ltd., the Sub-committee has been able to include an addendum devoted to surface defects in steel strip products.

The second edition is now available as **Special Report No. 63;** it consists of 62 quarto pages, illustrated with 134 excellent photographs of surface defects, and bound in cloth with stiff board covers. The Report costs £1 5*s*. 0*d*. (15*s*. 0*d*. to Members), and may be obtained on application to

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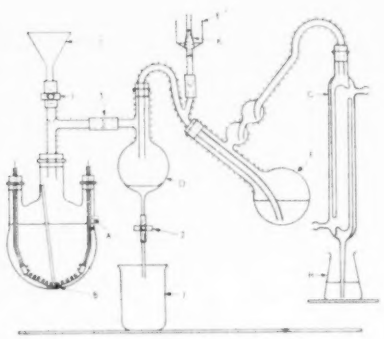
The Determination of Nitrogen in Steel



Great interest has always been shown in the role of nitrogen in determining the properties of ferrous alloys. Reliable and accurate methods are therefore necessary for determining the content of this element in alloys.

The establishment of the reproducibility and accuracy of analytical methods to serve as routine or reference procedures for the many alloys now produced is a major undertaking. It was for this reason that the Nitrogen Group of the Steelmaking Division of the British Iron and Steel Research Association was set up eight years ago. The Report of the Group has now been published as No. 62 in The Iron and Steel Institute's Special Report Series. The information it contains will be of inestimable value to all those interested in this subject.

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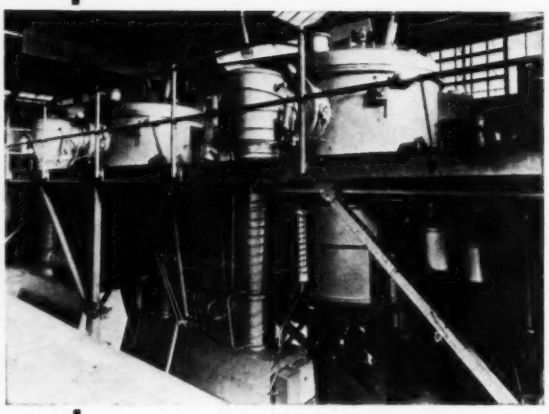

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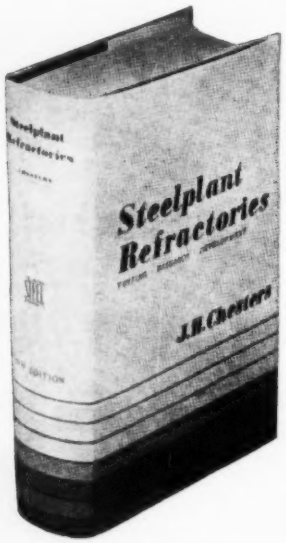

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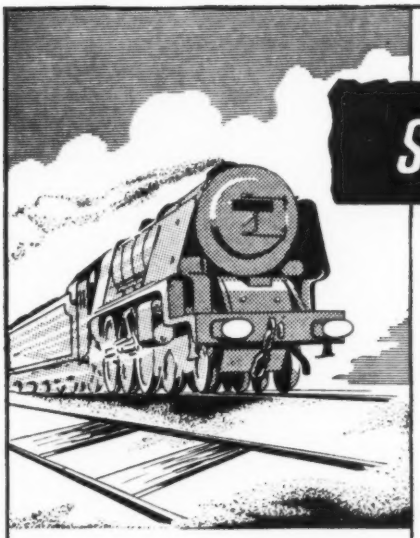

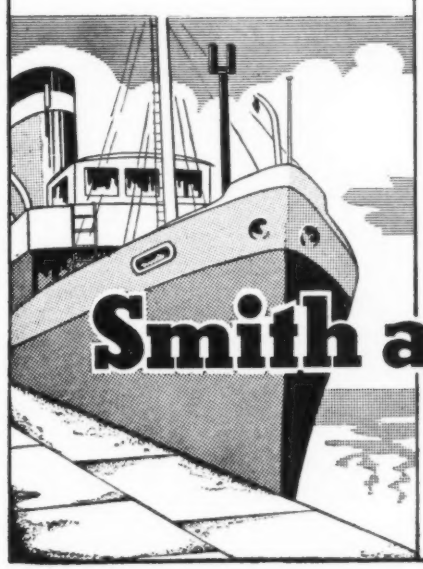

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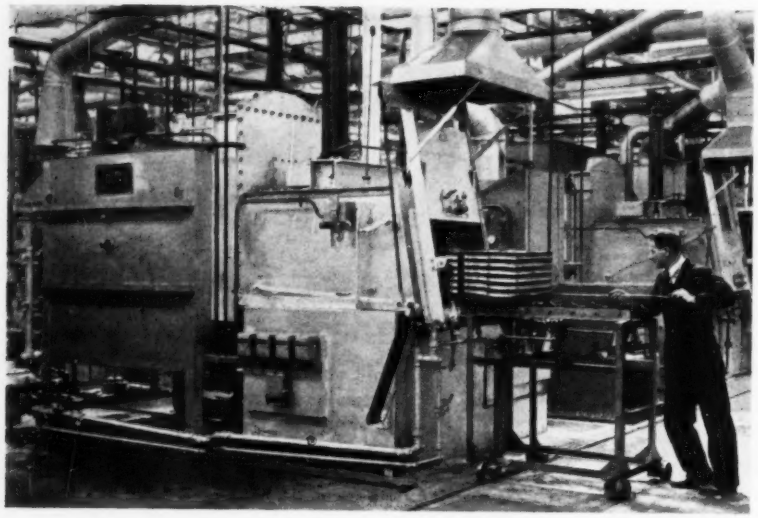

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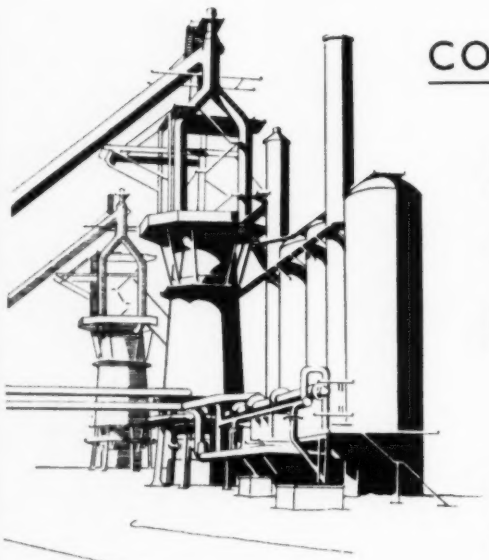

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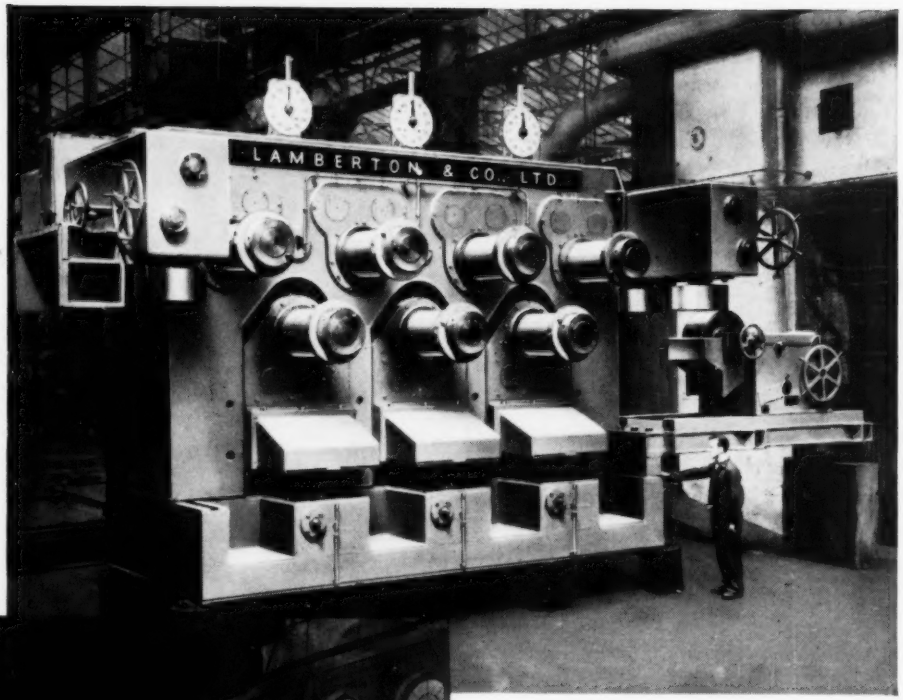
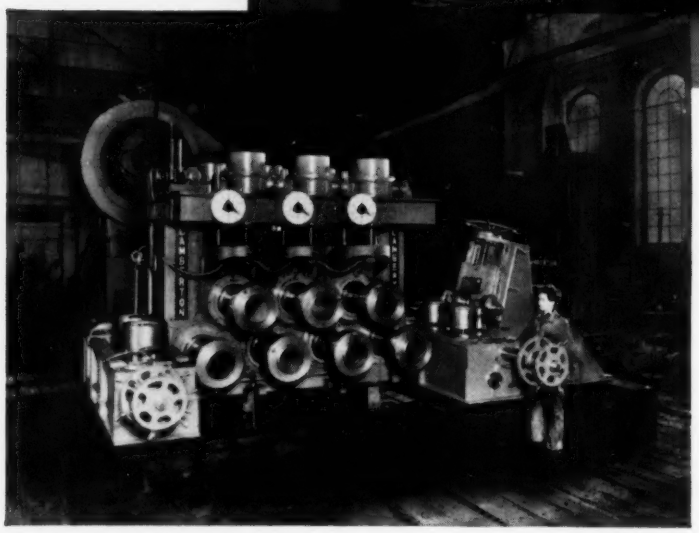


This 26 inch roller straightening machine is provided with driven vertical guide rollers. All-roller bearing spindles are used and the frame is of heavy welded steel construction. Its capacity is 13in. × 5in. joists and 40lb. per yard rails.



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